OCTOBER 1960
Volume 196 Part 2

Established 1869
Incorporated by Royal Charter 1899

Properties of an experimental Mo-V steel rotor forging
A. I. Smith et al. 117

Electrolytic production of sulphur-free iron
W. T. L. Neal et al. 131

Adherence of oxides on metals
R. F. Tylecote 135

Kinetics of the martensite transformation in iron alloys
W. S. Owen and A. Gilbert 142

Mechanical properties of some tempered alloy martensites
J. M. Capus and G. Mayer 149

Dislocations in α-iron
D. G. Brandon and J. Nutting 160

Effect of heat-treatment on high-temperature ductility
K. J. Irvine et al. 166

Ductility aspects of 18-12-1Nb steel
R. J. Truman and H. W. Kirkby 180

Heat-affected zone cracking in welded high-temperature austenitic steels
R. N. Younger and R. G. Baker 188

Failures in a blooming mill
F. E. Stacey 195

Lighting in the iron and steel industry
D. R. M. Nisbet 207

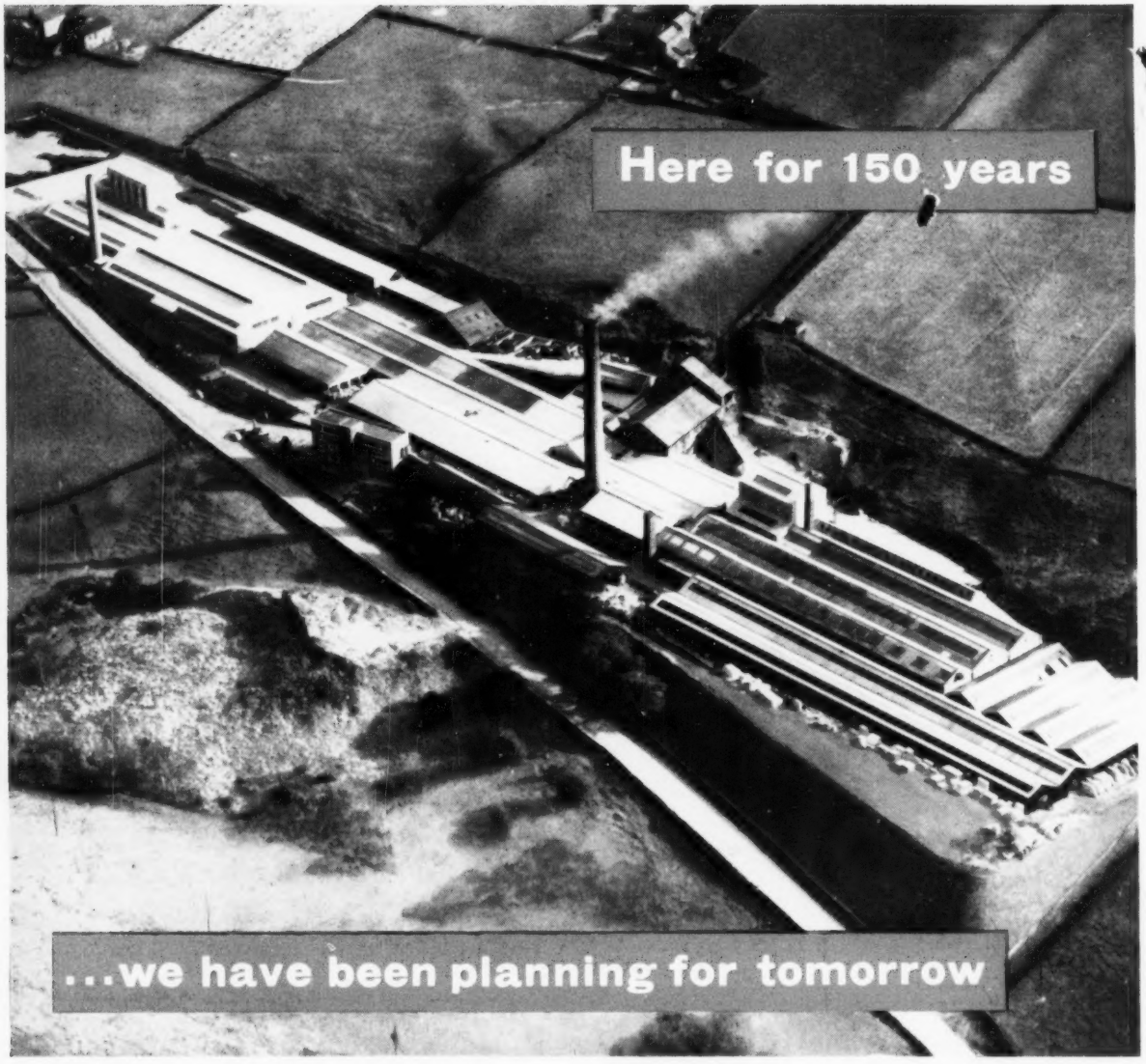
Engineering aspects of the Redbourn development scheme
P. M. Hesketh 213

Developments at Appleby-Frodingham since 1954
E. A. Atkin and E. F. Farrington 217

News 225 Abstracts 229

A full list of contents is given on the Title Page (see page 63 of the Advertisement Section)

JOURNAL OF THE IRON AND STEEL INSTITUTE



for increasing steel output

using

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Inside front cover

The latest development in Open Hearth Steel-making Practice

WELLMAN

Britain's foremost builders of Steel-making Plant

now offer the

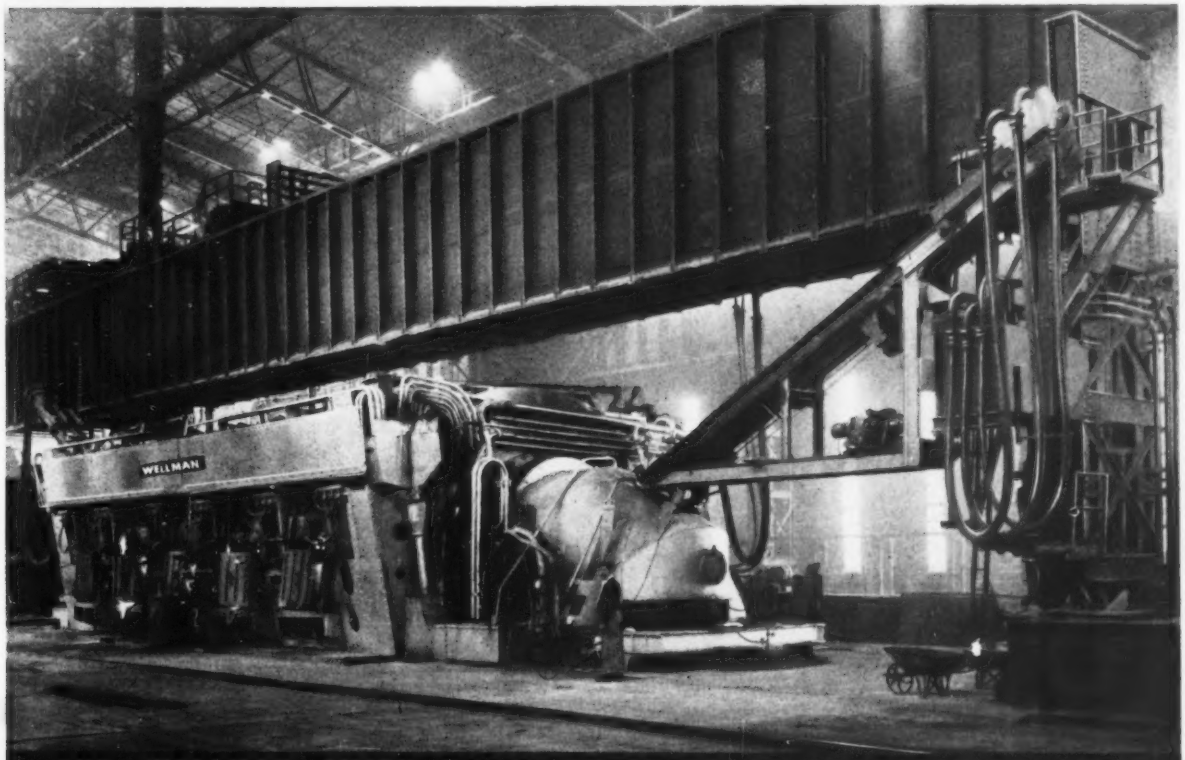
AJAX Oxygen Steel-making Operation

This new steel producing technique has been incorporated in the furnace illustrated. It was invented and developed by Appleby-Frodingham Steel Company, and has already produced over ¼ million tons of ingots. A second unit has now been commissioned.

Outputs have been greatly increased with an average of five thousand tons of steel per week.

This Furnace has now 90% availability together with lower than normal refractory costs.

As licensees we invite you to consult us for either converting existing plant or providing new furnaces incorporating the Ajax Oxygen Steel-making Operation.



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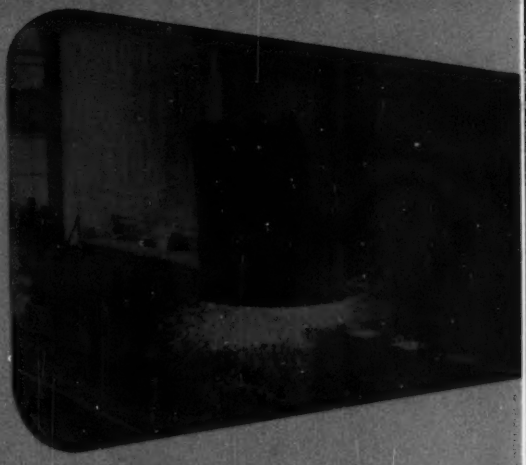
BRITISH STEEL AT ITS BEST



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Normanby Park Steel Works,
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GKN

Modern carving of a sow and litter in All Saints Church, Beighton. Photograph by Ruth Meyler & F. C. Sillar

50 MILLION PIGS GO TO MARKET!

In the early days of the industrial revolution manufacturers frequently looked for familiar and easily recognisable descriptions of their products. To the countryman transplanted into industrial surroundings nothing was more familiar than a pig; a fact which is emphasised by carvings in scores of old churches depicting pigs suckling their litter, rooting for acorns and playing the bagpipes. (Apparently this last pastime was forsworn by pigs many years ago because of unpleasant associations between bagpipes and pigs bladders.) Consequently, when iron from Blast Furnaces was directed into long troughs and from there into short and narrow troughs placed at right angles, a resemblance to sows suckling their appendaged litter led to the description of 'pig iron'.

This series of advertisements illustrates some of the 'ecclesiastical' pigs and emphasises the 50,000,000 'special' pigs which Warner and Company have supplied to Iron Foundries producing High Duty Castings.

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"Pioneers of Refined Pigiron"

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WARNER & CO. LTD., MIDDLESBROUGH

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of the specialised television in the iron and steel industry is

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There are four sound engineering reasons why

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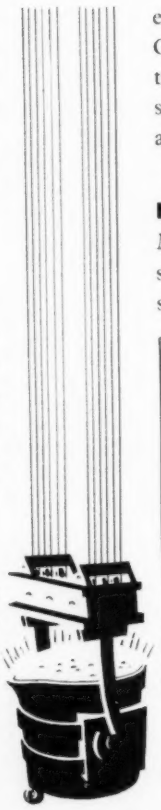
When an equipment needs servicing, it needs it immediately if the plant in which it is integrated is to continue to operate profitably. In addition to an efficient after-sales service, a special television equipment maintenance course is provided at Marconi College, Chelmsford, and customers are strongly urged to send those of their personnel who will be responsible for the running of their television systems to attend this.



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in providing an engineering service to the Metal Industries

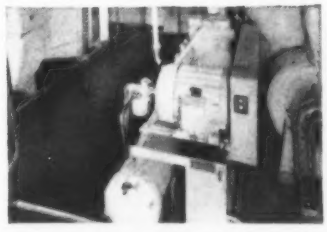
SOLANO

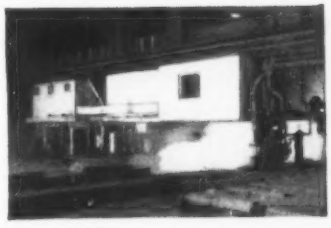
—the bye-word

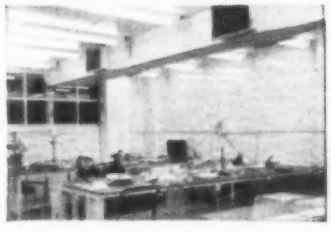
. . . whenever experience, skill and efficiency in engineering are discussed by management

The Solano Organisation — offering a broad and comprehensive service backed by a highly qualified technical staff — brings to industry a wealth of experience and 'know how' which is unparalleled in many spheres of engineering. Requiring experts in the design, fabrication and erection of many types of engineering projects, leading names in industry have sought the down-to-earth advice, close assistance and ultimate benefits of the Solano Organisation. Today, the name Solano stands supreme for advice and assistance on Air Conditioning, Heating, Ventilation, Dust and Fume Extraction, Light and Medium Structural Steelwork.

THE SOLANO ORGANISATION IS EQUIPPED TO DESIGN AND ERECT CONTROL PULPITS AND CRANE CABS, FULLY AIR CONDITIONED AGAINST HIGH AMBIENT TEMPERATURES AND READY FOR SERVICE.







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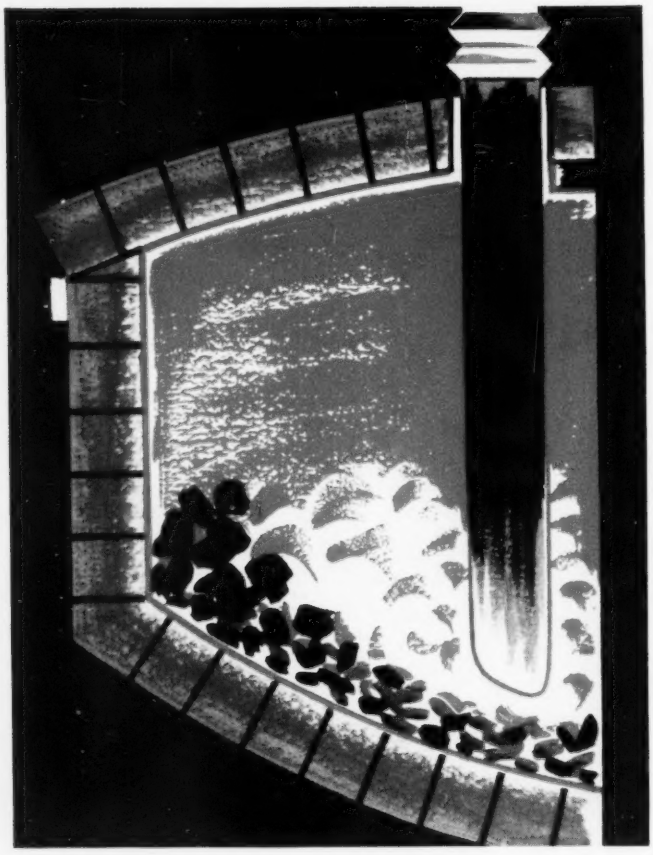
The Lancashire Steel Manufacturing Co., Ltd.
Motherwell Bridge & Engineering Co., Ltd.
The Park Gate Iron & Steel Co., Ltd.
Stewarts and Lloyds Ltd.
Taylor Bros. & Co., Ltd.
The United Steel Companies Ltd.

rdh 6811



why ACHESON TRADE MARK graphite electrodes

Give You Dependable Service!

"Acheson" Graphite Electrodes are manufactured from specially selected high grade raw materials which are carefully processed at all stages of manufacture under controlled conditions to give a final electrode for your arc furnaces with these properties at their optimum—

LOW RESISTANCE

HIGH MECHANICAL STRENGTH AT OPERATING TEMPERATURES

EXCEPTIONAL THERMAL SHOCK RESISTANCE

"Acheson" Graphite Electrodes are made in the United Kingdom only by

BRITISH ACHESON ELECTRODES LIMITED of SHEFFIELD.

ACHESON

TRADE MARK

GRAPHITE ELECTRODES

The term ACHESON is a registered trade mark

HOLSET

CHOSEN BY
THE UNITED STEEL COMPANIES LIMITED

FLEXIBLE RUBBER BLOCK COUPLINGS

Appleby-Frodingham Steel Company
Samuel Fox & Company Limited
Steel Peech & Tozer
Workington Iron & Steel Company

CHARACTERISTICS OF HOLSET COUPLINGS

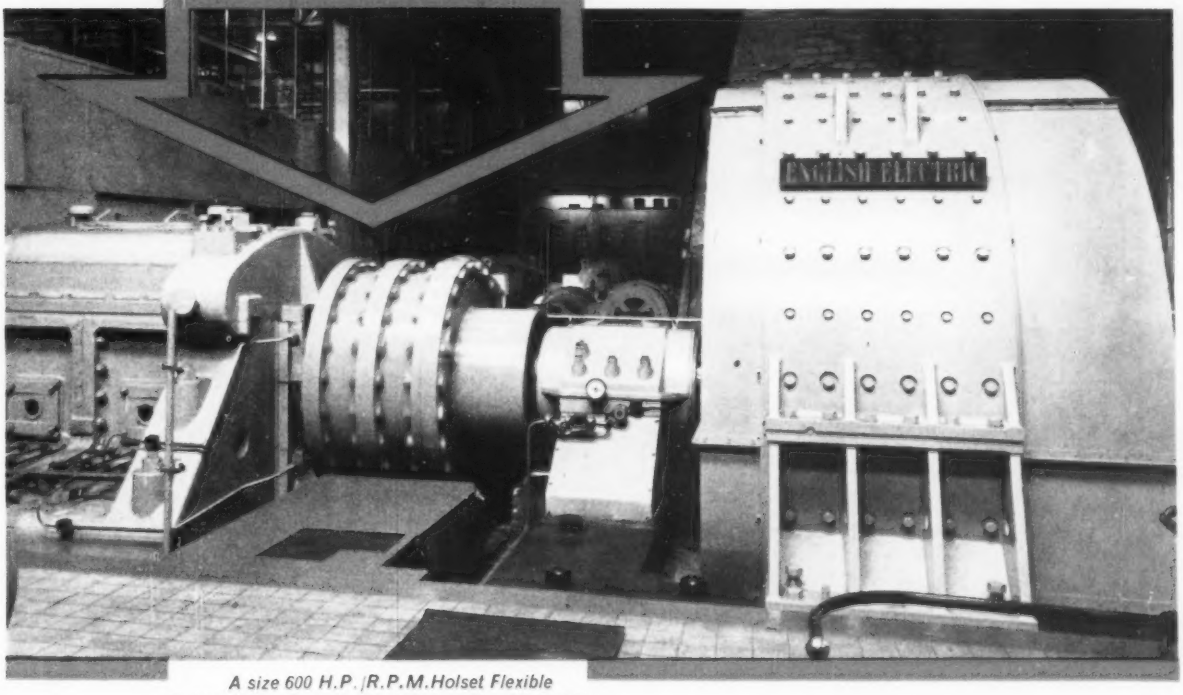
Can accommodate limited angular and parallel misalignment.

Absorb shocks and operate noiselessly under any conditions.

The multi-row arrangement of the blocks results in a small diameter and, therefore, a lower WR^2 value than any other torsionally resilient coupling.

No end thrust on the bearings of the connected shafts.

No lubrication or maintenance required.



A size 600 H.P. /R.P.M. Holset Flexible Rubber Block Coupling installed on the 40" Slabbing Mill at the Stocksbridge Works of Messrs. Samuel Fox & Company Limited

THE HOLSET ENGINEERING CO. LTD
TURNBRIDGE, HUDDERSFIELD.
TELEPHONE HUDDERSFIELD 7480
TELEGRAMS HOLSET, HUDDERSFIELD

Bulk Log Handling

Mitchell conveyors at Bowaters' Mersey Division carry 100 tons of logs an hour...

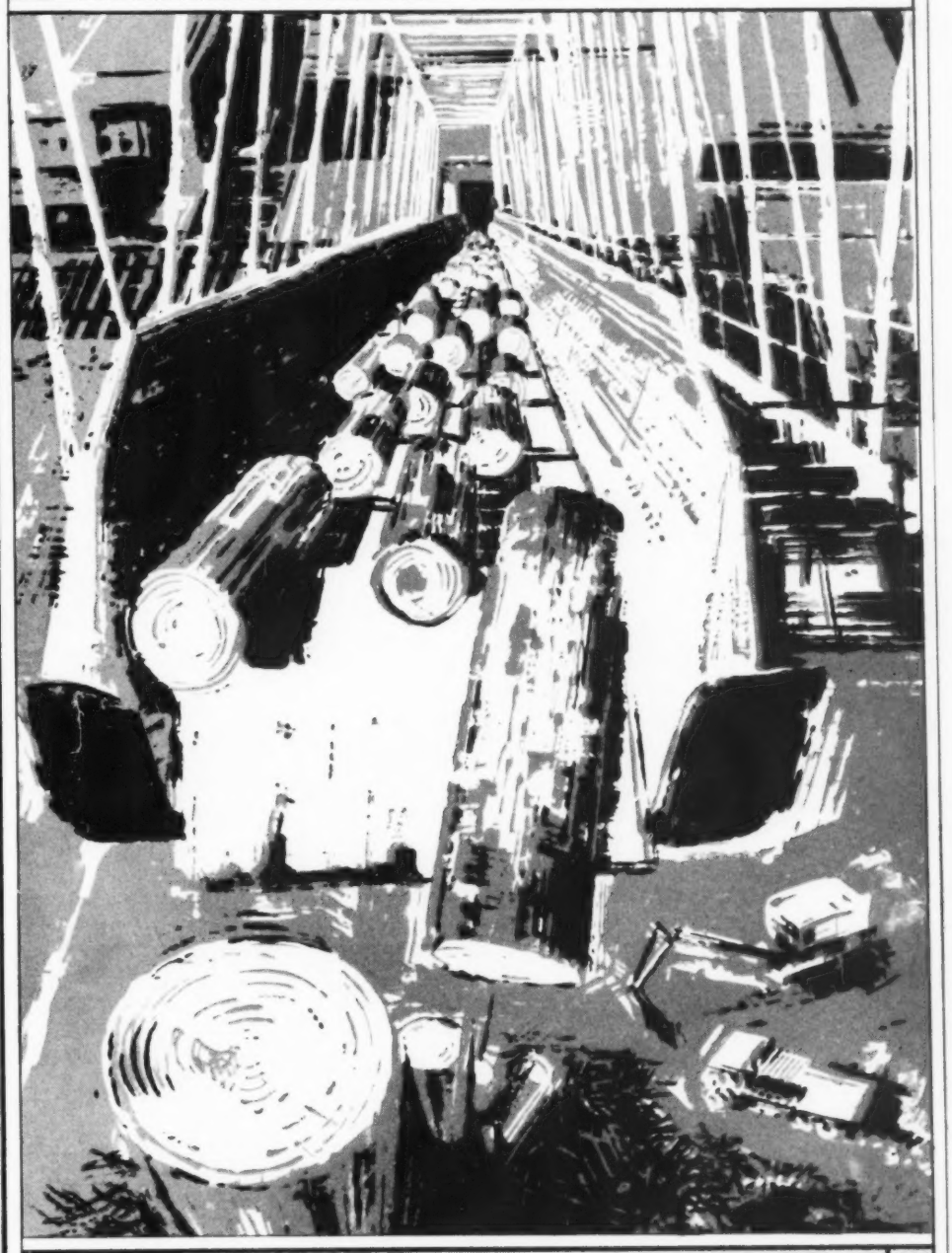
The problem of conveying and stacking logs in really large quantities is comparatively new to this country. Yet Mitchell Engineering's recently completed external conveyor system at Bowaters' Mersey Division handles over 100 tons of logs an hour—and more!

This vast capacity is attained by an efficient arrangement of chain and belt conveyors, 1,250 ft in total length, which carry logs either straight to the pulping mill or to a storage pile. The logs carried straight to the mill pass through a debarking plant and provision is made for re-circulating any log not properly debarked. Logs for the storage pile are carried up a 120 ft mobile stacker.

Mitchells are active in the field of boiler design, cableways and ropeways, wagon handling, transporters and conveyors, cranes, mills, ship loading plant and nuclear power.

Mitchell Engineering Group Limited | *One Bedford Square London WC1*

The mobile stacker seen in operation here has already handled over 146,000 tons of logs



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To meet the ever growing demands for....

....Feedex and other exothermic materials, we are building new extensions to our factory—the completion of these extensions will guarantee your future requirements

Foseco

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Drayton Manor, Tamworth, Staffs.

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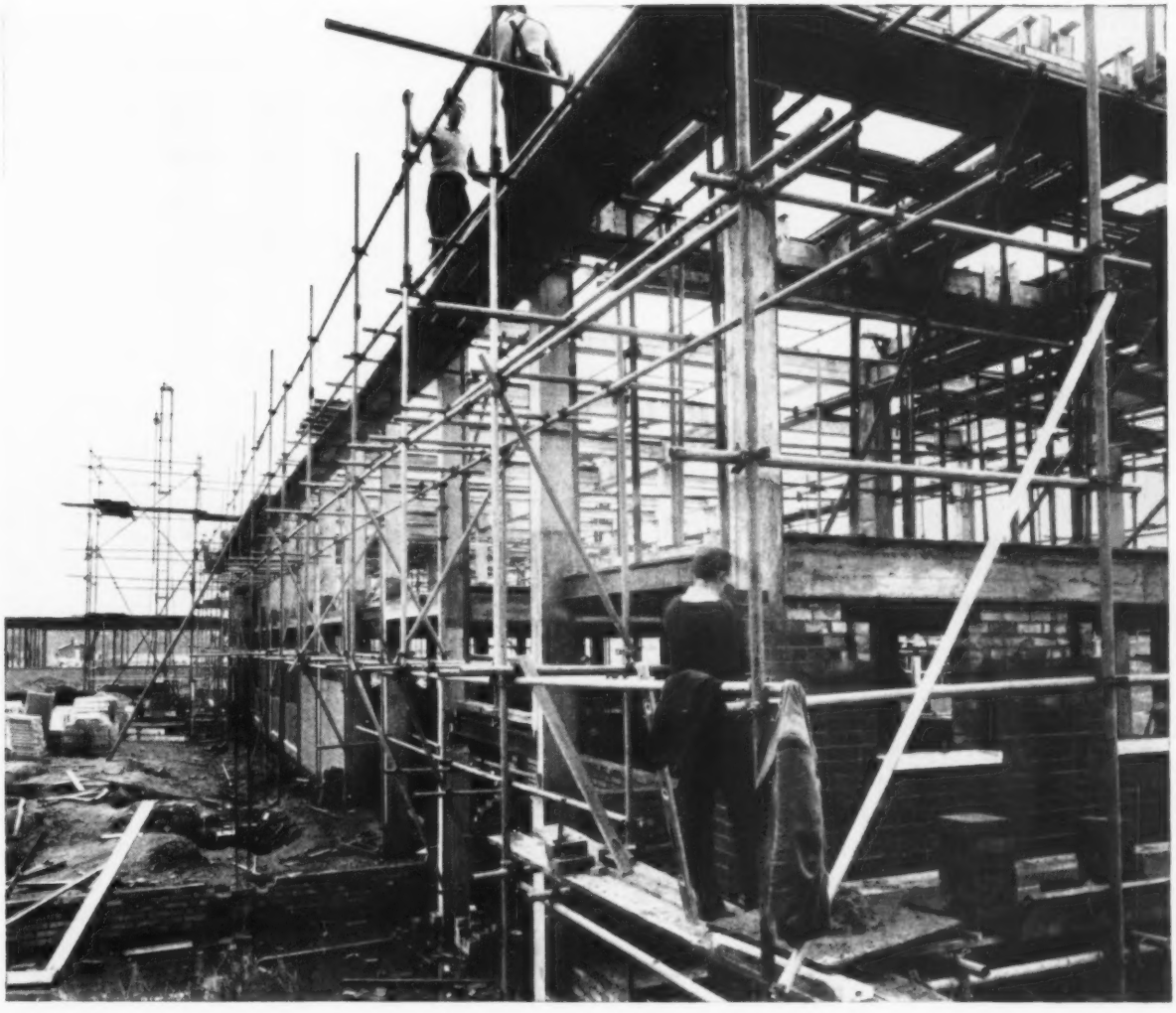

The best electric motors are the ones you can forget, motors that will operate under the most arduous conditions without giving trouble, in fact Class H motors. Class H—silicone insulated—motors withstand frequent overloads, heat and humidity to a vastly greater extent than Class A or B machines. Continuous running tests have proved that the insulation life of Class H motors is ten times *as long as that of their Class A or B counterparts. And don't forget, if you're going in for rewinds make them SILICONE rewinds! That's the painless way to convert those troublesome motors into consistent performers.*

WHERE THE GOING IS TOUGH

Silicones strengthen the weak link in all forms of electrical equipment—its insulation. Write to the nearest branch of Midland Silicones for the names of rewinders experienced in the use of silicone insulation.

MS

MIDLAND SILICONES LTD

(Associated with Albright & Wilson Ltd. and Dow Corning Corporation)

first *in British Silicones*

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MSE6

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and Spares are available ex stock

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18 Sandyford Place, Glasgow, C.3
Telephone: Glasgow, City 3951

LEEDS
Brotherton Chambers, Grace Street, Leeds 1
Telephone: Leeds 33481

LONDON
32 Old Queen Street, Westminster, London SW1
Telephone: Whitehall 3875

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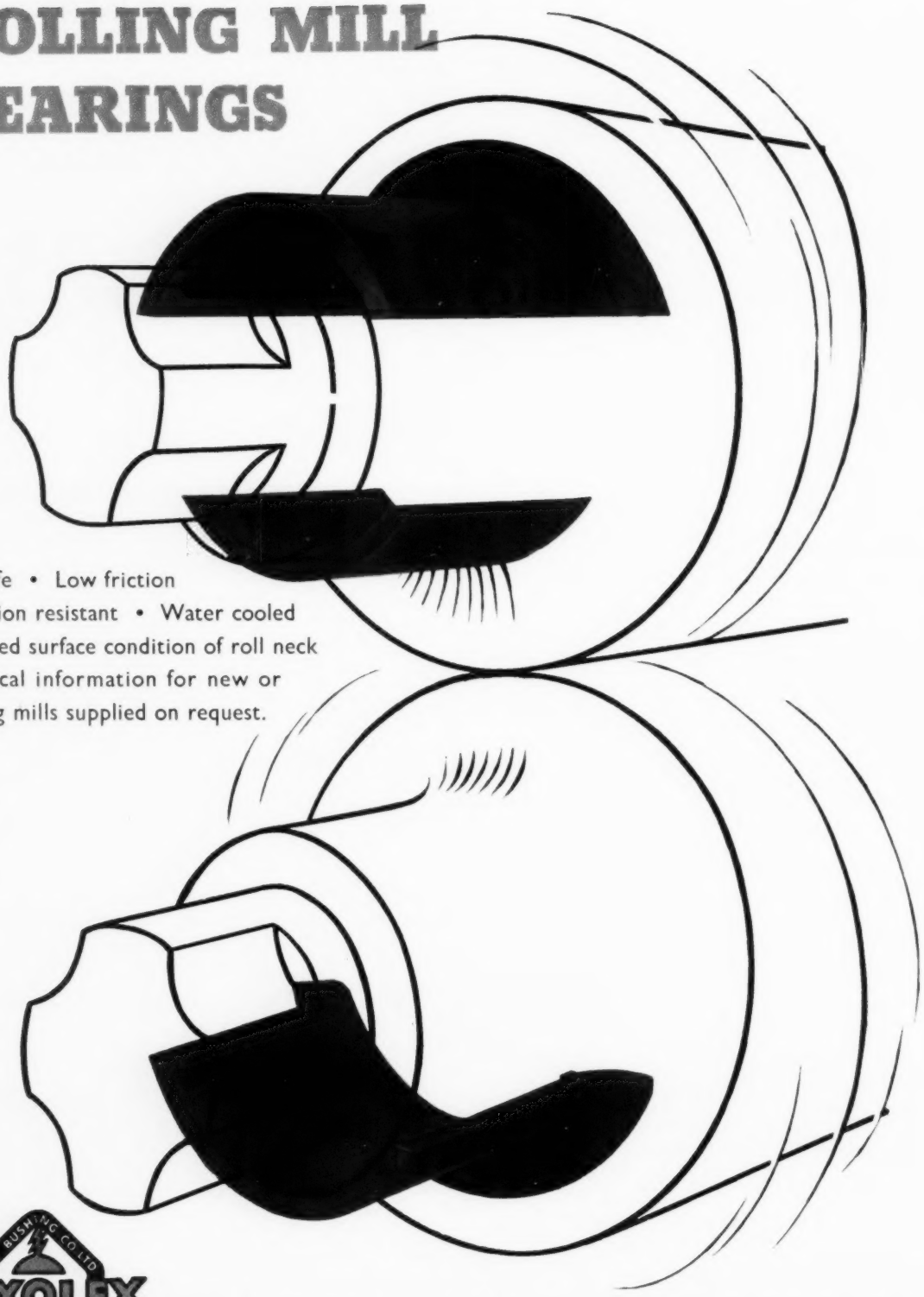

Long life • Low friction
Corrosion resistant • Water cooled
Improved surface condition of roll neck
Technical information for new or existing mills supplied on request.

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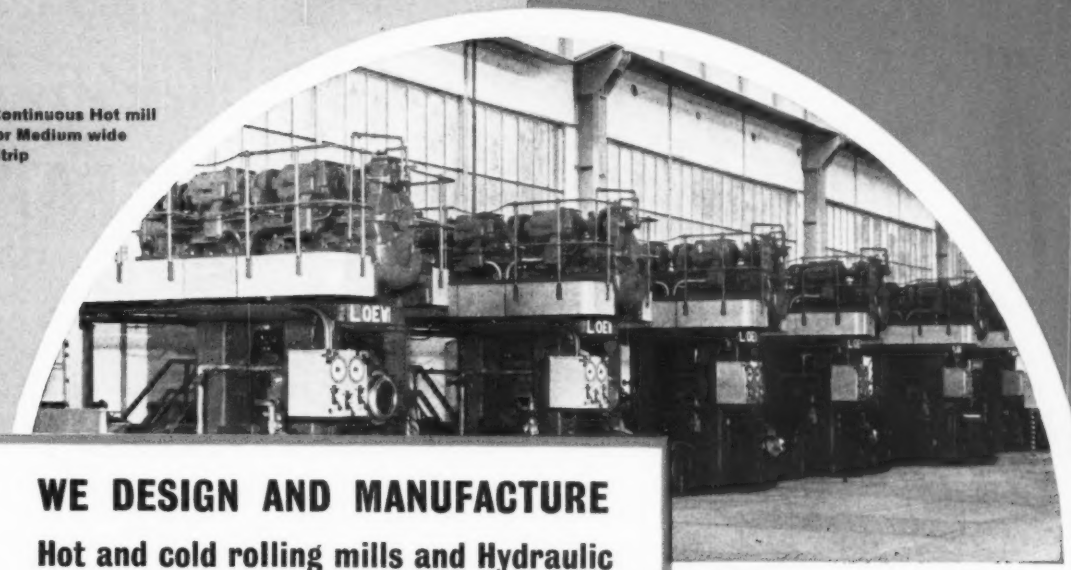
ROLLING MILLS



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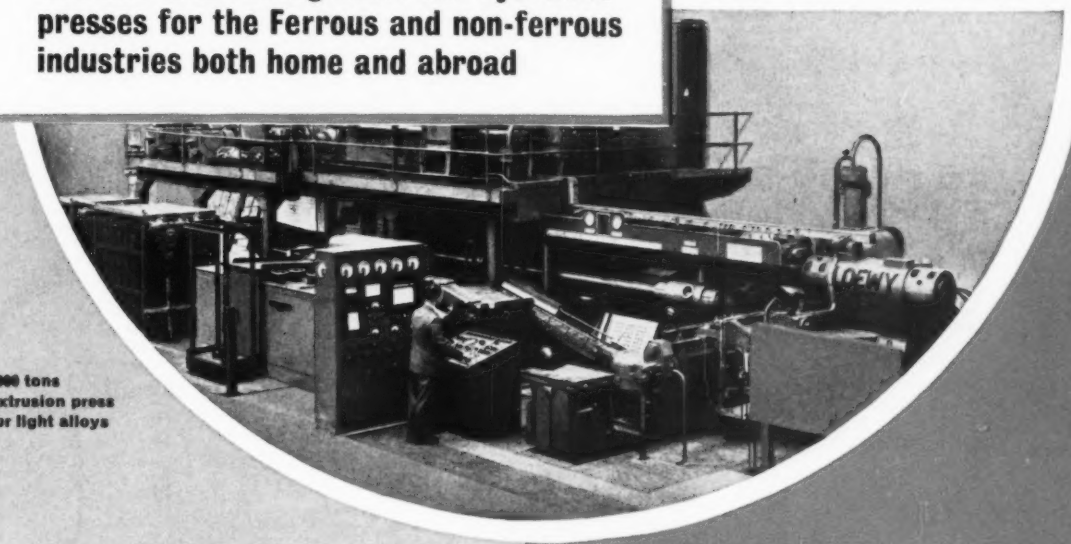
Hot and cold rolling mills and Hydraulic presses for the Ferrous and non-ferrous industries both home and abroad



2000 tons extrusion press for light alloys

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THE LOEWY REG'D. TRADE MARK ENGINEERING COMPANY LTD.
BOURNEMOUTH ENGLAND

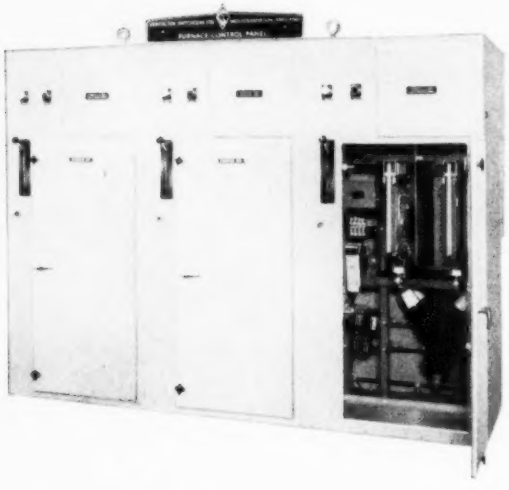
Electric Furnace

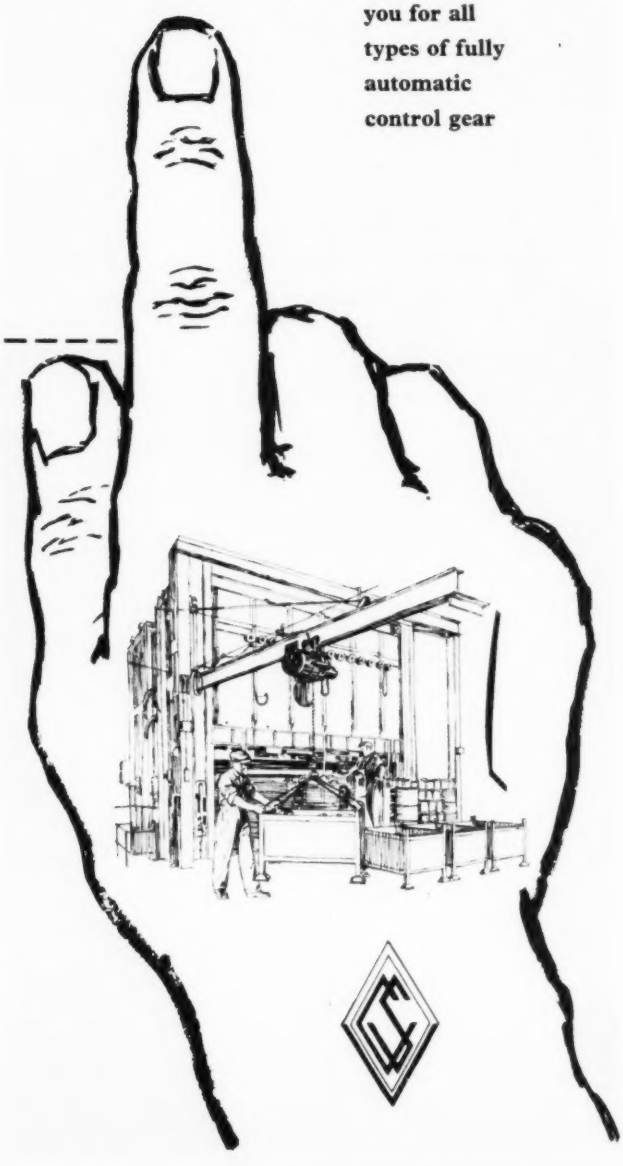
Automatic CONTROL GEAR

The panel illustrated is a three panel 6,600 volt 3 phase, 50 cycle switchboard controlling induction Heating furnaces. Our system of Switchboard construction provides complete flexibility and adaptability for designing and manufacturing control gear for all types of electric furnaces and steelworks.

Send for Publication No. 121/0660.

May we quote you for all types of fully automatic control gear





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Telephone: Wolverhampton 25911/7

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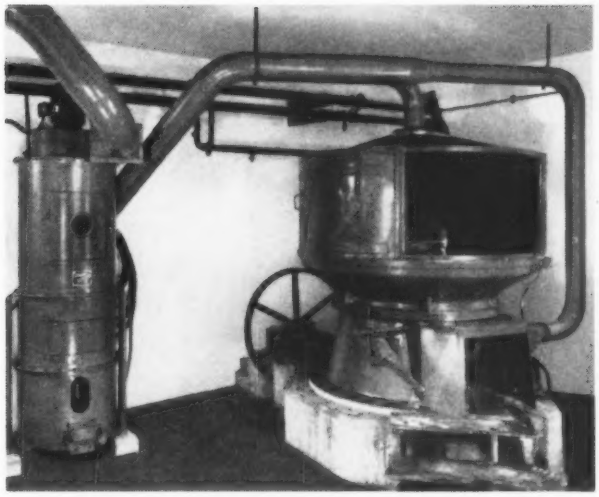
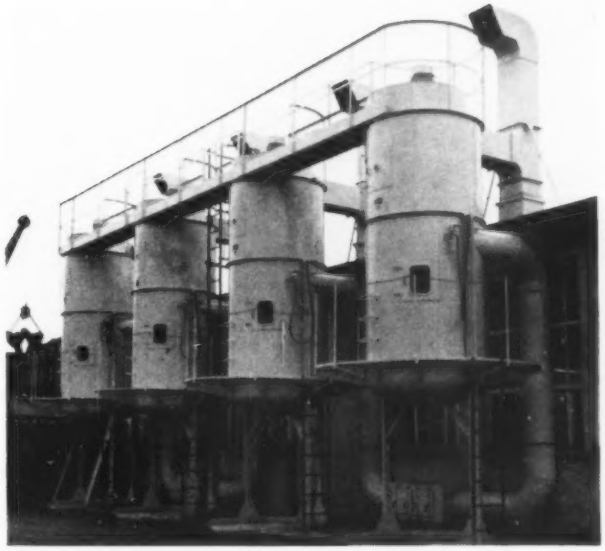
HOW CAN YOU BEST HANDLE YOUR DUST?



Wet or dry? If wet, or when it is simpler and more advisable to dispose of "valueless" dust in sludge form, the VISCO-HANDTE Dust Collector is the most economical way of collecting dust from foundry knockouts, dressing, fettling, ore handling and subsidiary operations. It is manufactured and sold exclusively in the United Kingdom by VISCO. Enquiries are invited to THE VISCO ENGINEERING CO. LTD., Stafford Road, CROYDON, SURREY. Croydon 4181.



VISCO-HANDTE

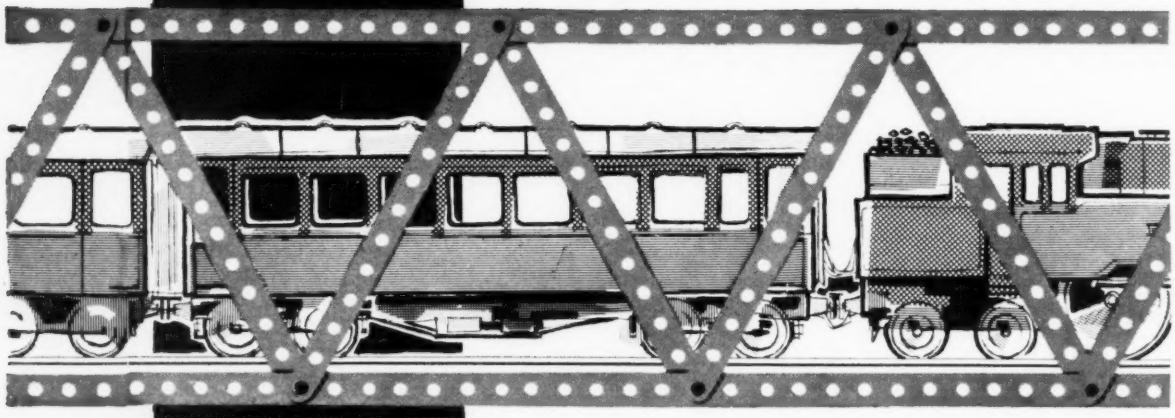
Wet type Dust Collectors

COAL keeps MECCANO

There are some names we learn in our boyhood that never lose their magic—and Meccano is one of them. Today, this name, so intimately connected with enamelled pieces of steel strips and plates that made every youngster a constructional engineer has an even wider appeal. Together with Hornby trains—electric and clockwork—Dinky Toys and Bayko, Meccano now provides a quartet of delight for boys and girls—and their fathers!

The far-sighted management that has kept Meccano a best-seller shows the same sound judgment when

PROGRESSIVE INDUSTRY IS GOING

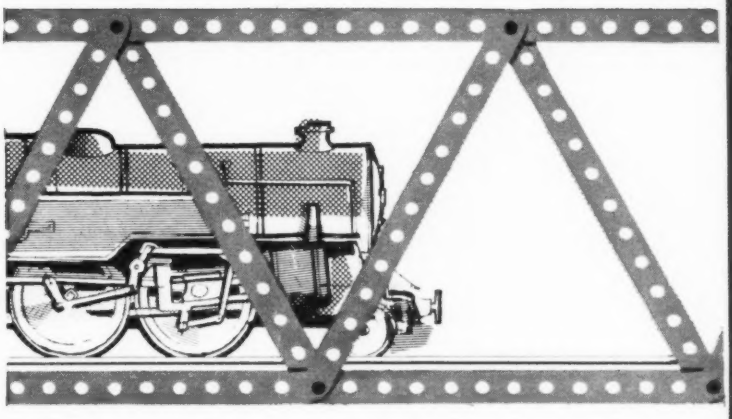


running like clockwork

dealing with grown-up affairs, like choosing *coal* to fuel its great powerhouses at Liverpool and Aintree. Coal was chosen because coal provides more heat for the money than any other fuel. Coal, mechanically stoked, is smokeless. And in our coalfields there is enough coal to keep British industry going at top production for generations to come.

When you are required to make a decision about fuel remember Meccano. Coal has helped them to build up a model industry. Choose the same fuel—and you've made the most constructive start.

FORWARD ON COAL



ISSUED BY THE NATIONAL COAL BOARD

Coal Provides the Power at Meccano's factory at Binns Road, Liverpool

Mr. Gibbs, Chief Engineer, comments:

"We at Meccano operate all power and steam process equipment from coal.

The present boiler plant consists of two water tube boilers and economisers fitted with usual mountings, including feed water regulators. These boilers were installed in 1928 and still give us a first-class service today. Their efficiency is high.

Since the original installation we have continued to make improvements. In 1948 further instruments were added, such as CO_2 recorders and indicators and flue gas temperature recorders. A feed water meter was installed in 1955 and soon afterwards a smoke indicator and recorder. More recently the reciprocating feed pumps have been superseded by electrically operated centrifugal pumps.

In this way we have kept pace with the increasing demands of production and made the best use of all our equipment.

Firing is by chain grate stokers and all fuel is handled mechanically. Steam is used for space heating and process work.

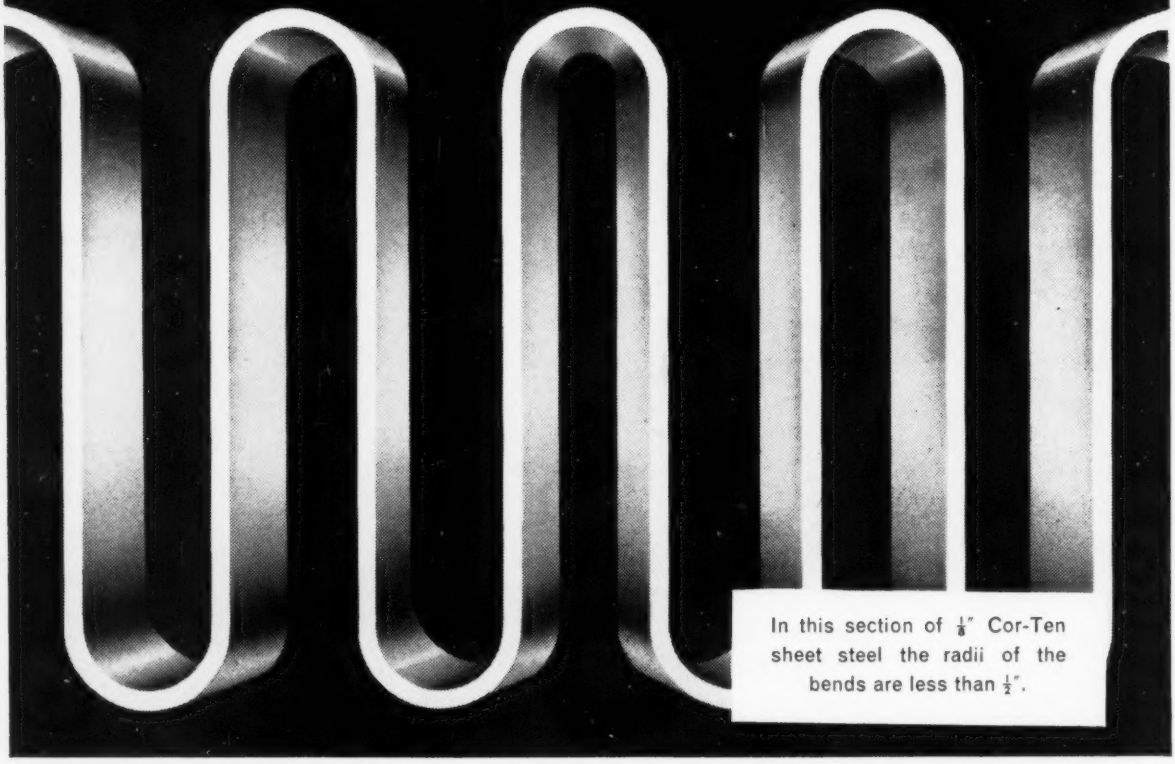
Each boiler is opened up for cleaning and inspection once a year. The flues are easily cleaned with a compressed air line, and on completion the boilers work for the following year with practically no maintenance.

Coal is obtained from a local colliery and deliveries are arranged daily to suit our requirements. The quality of the coal is always consistent and we are confident that we shall continue to use coal for years to come."

Here are some key facts and figures about the consumption of coal at Meccano:

Number of boilers: ..... 2 water tube
Method of firing: ..... chain grate stokers
Steam pressure: ..... 120 p.s.i.
Continuous max. rating: ... 6,500 lbs. per hour
Steam temperature: ..... 350° F.
Feed temperature: ..... 180° F.
Annual fuel consumption: ... 2,260 tons of coal

UNION CARBIDE CO-OPERATION WITH THE STEEL COMPANY OF WALES LIMITED MADE THIS POSSIBLE



In this section of $\frac{1}{8}$" Cor-Ten sheet steel the radii of the bends are less than $\frac{1}{2}$".

The Alloys Development Department of Union Carbide Limited, London, in their continuous search for new ways to make better steel, always work in close co-operation with the steelmakers themselves. They gratefully acknowledge the help of The Steel Company of Wales Limited who made it possible to run tests on the latest Union Carbide alloy additive—SILCHROMSOL. SILCHROMSOL, a new silicon chrome alloy additive, is now used in the production of SCW COR-TEN* high strength low alloy steel.

Photomicrographs show that the intermetallic compounds present when the steel is made with SILCHROMSOL are of a spherical shape instead of the "stringer" type which are associated with a lack of ductility. Tests indicate that SILCHROMSOL gives this type of steel vastly improved ductility. The new development has been well received and The Steel Company of Wales Limited have great confidence in the improvements which will result.

* SCW COR-TEN *steel is made in this country under licence from the United States Steel Corporation.*

CO-OPERATING WITH INDUSTRY

UNION CARBIDE

UNION CARBIDE LIMITED · ALLOYS DIVISION · 8 GRAFTON STREET · LONDON, W.1 · MAYFAIR 8100

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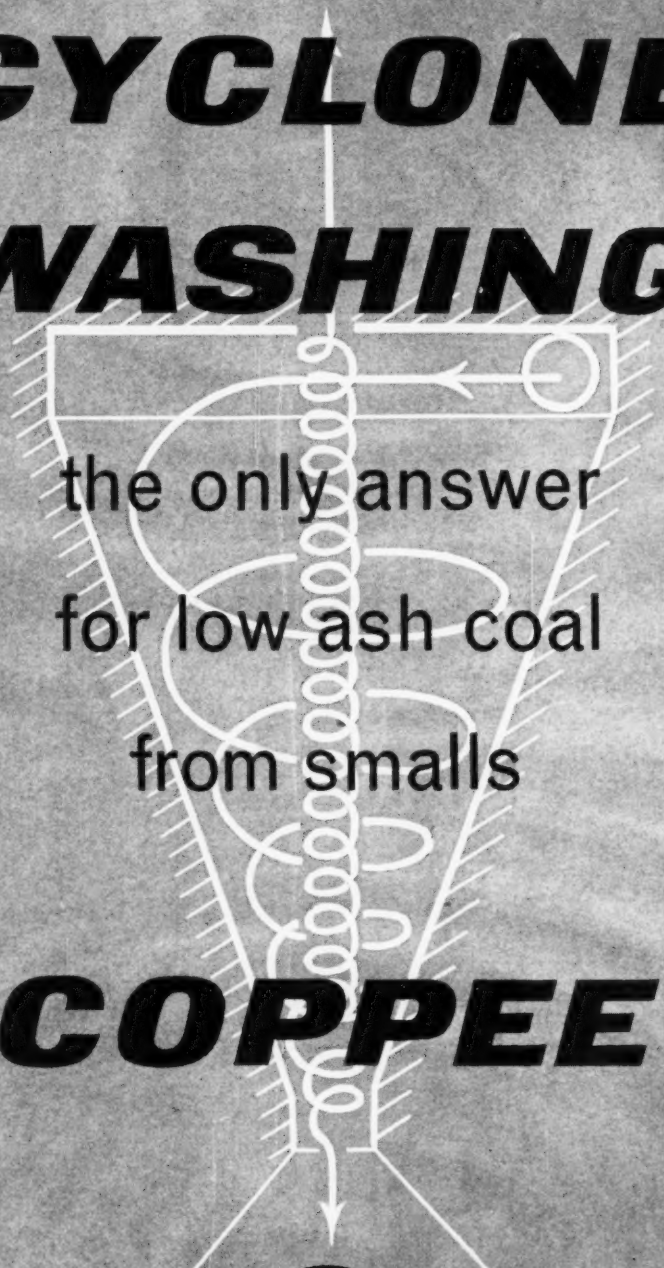

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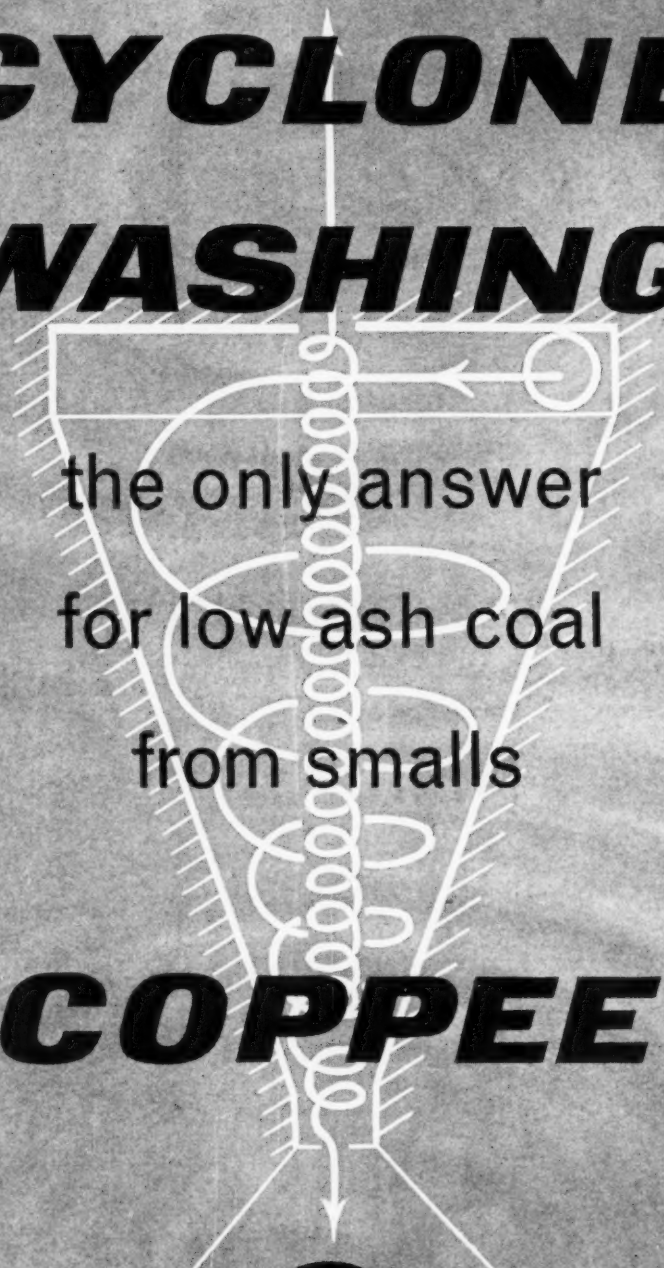

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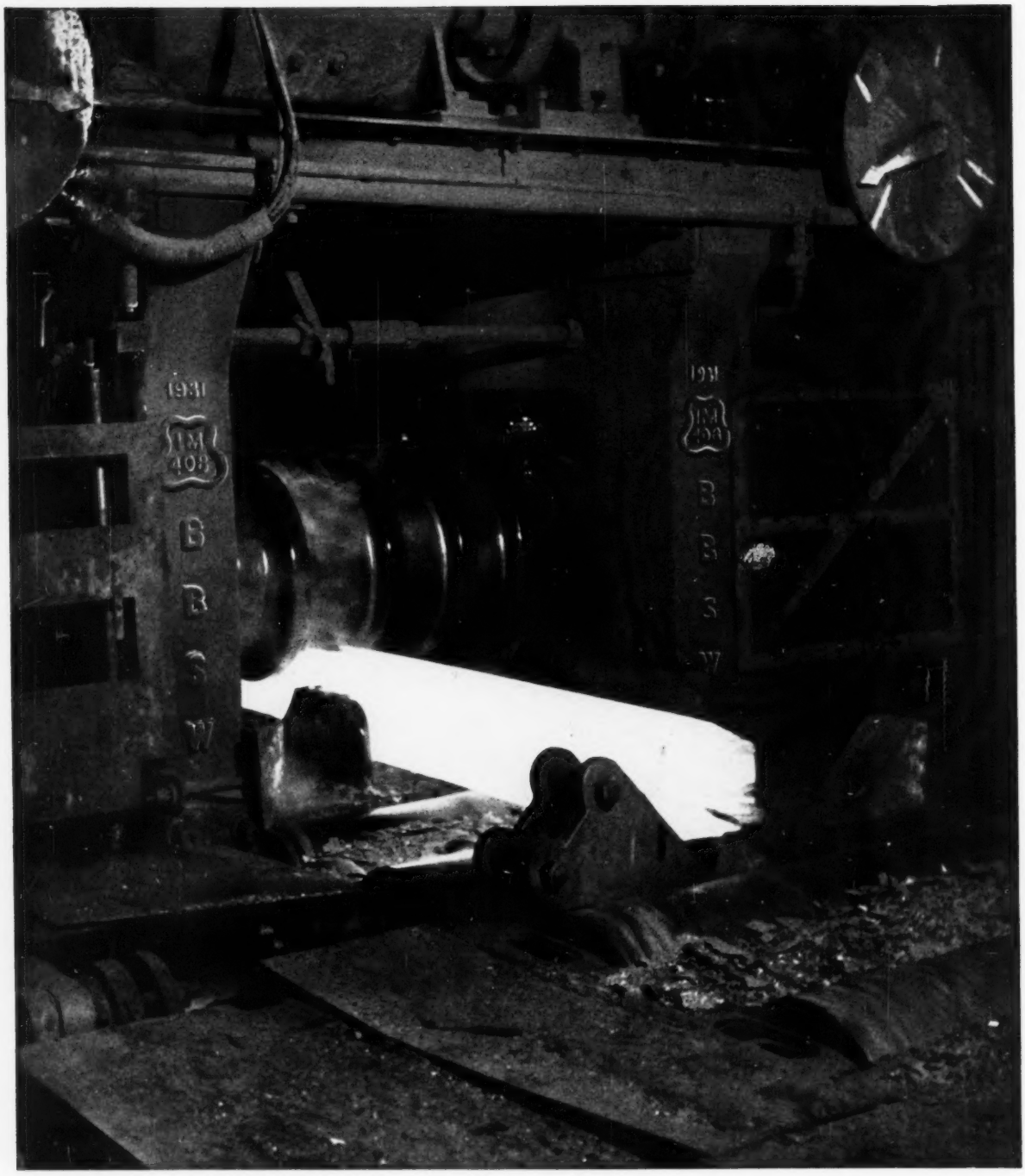

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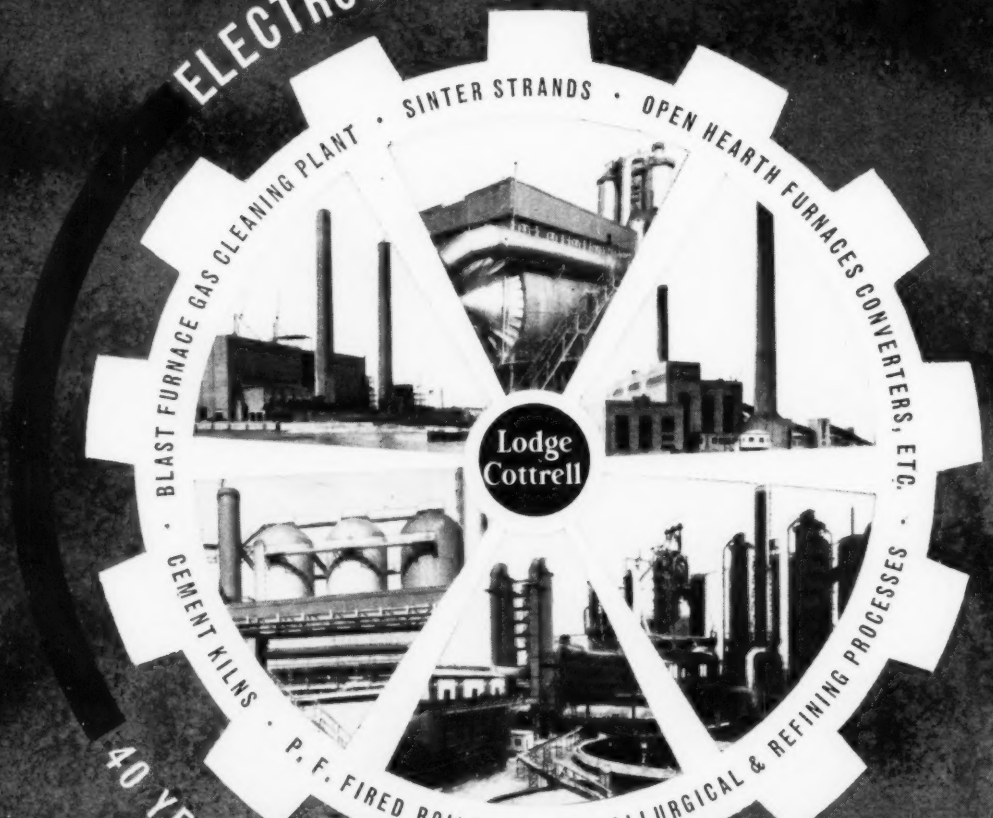

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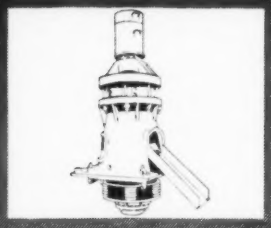

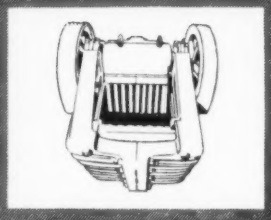

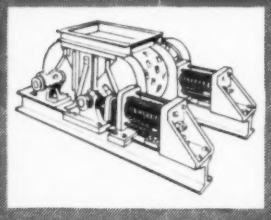

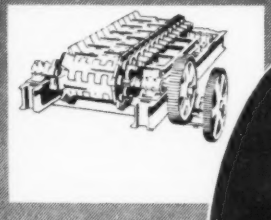

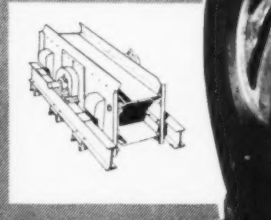

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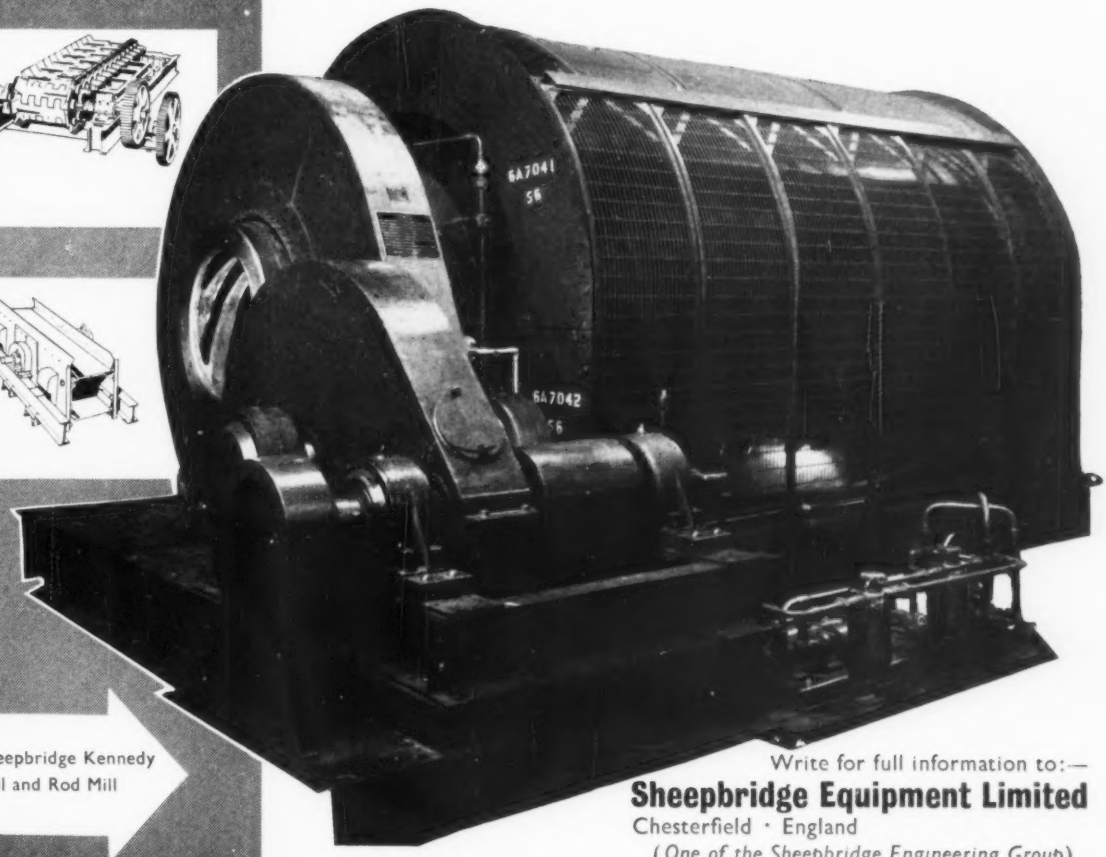

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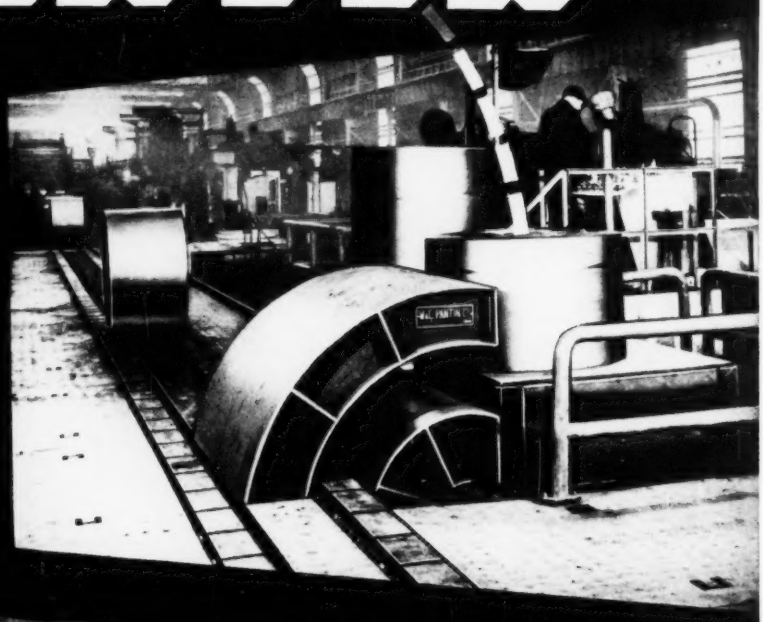

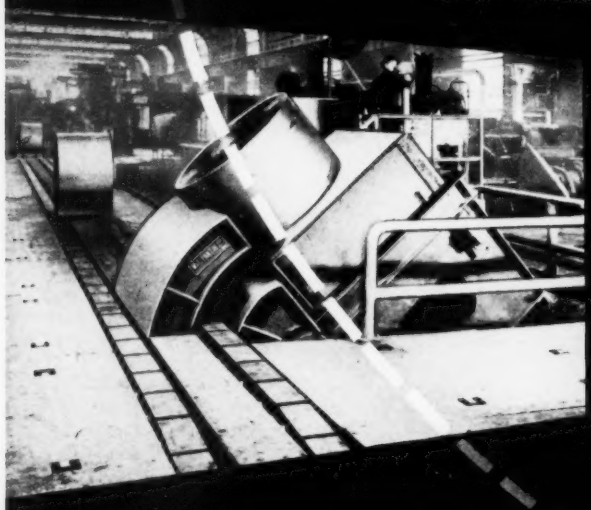

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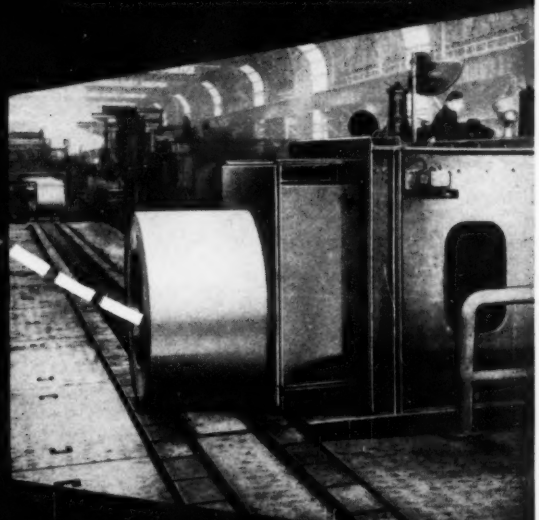

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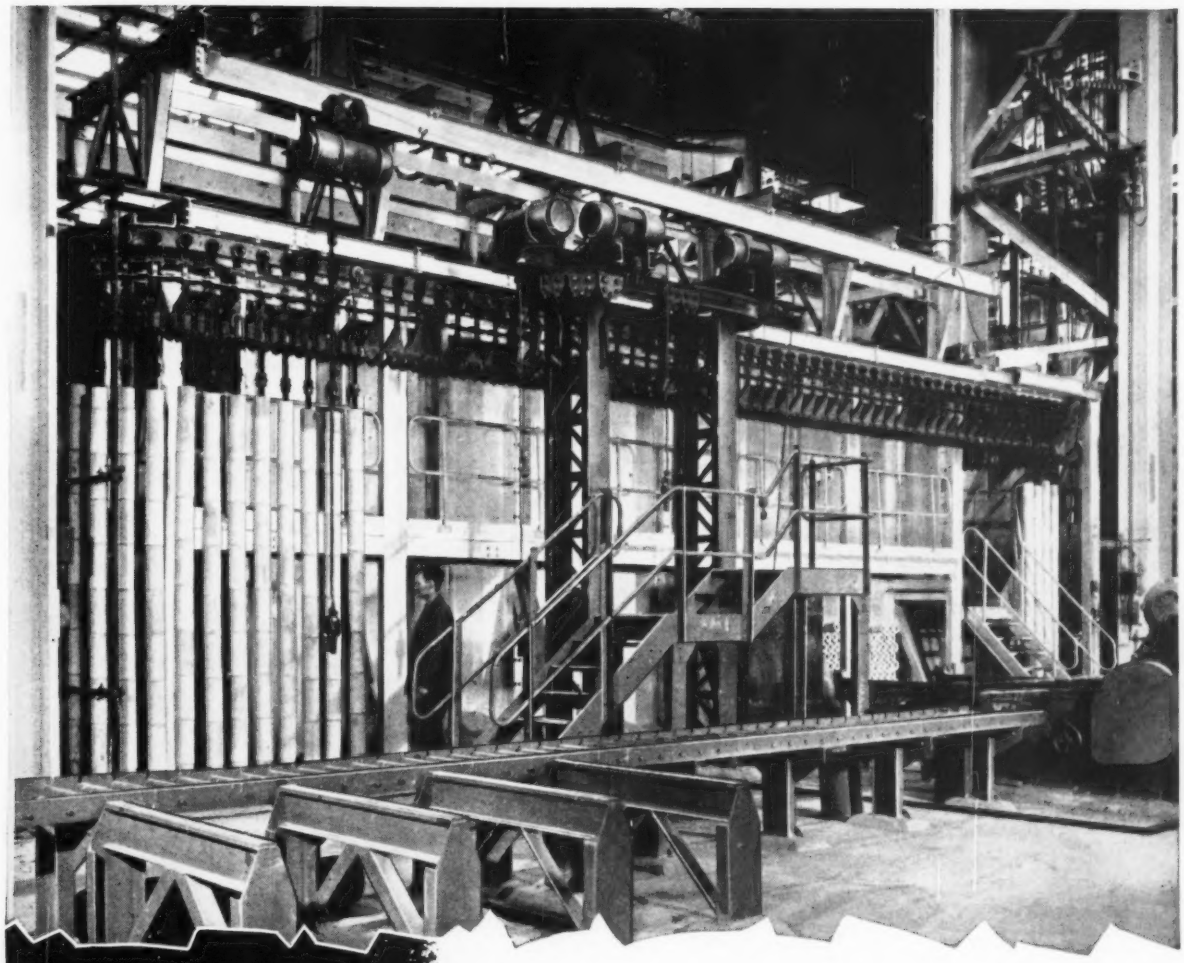

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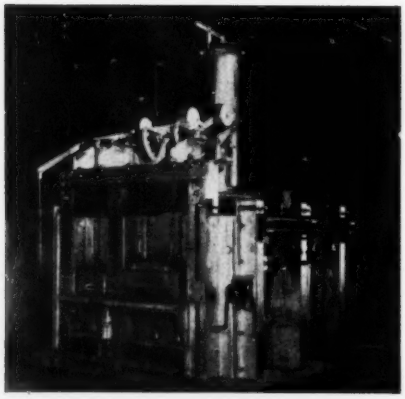

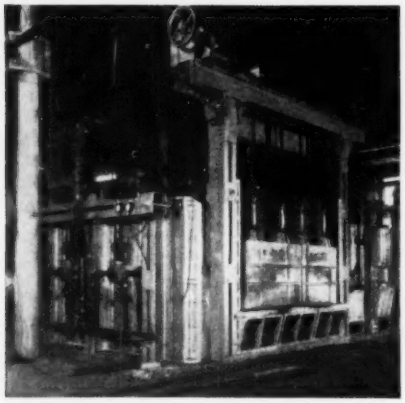

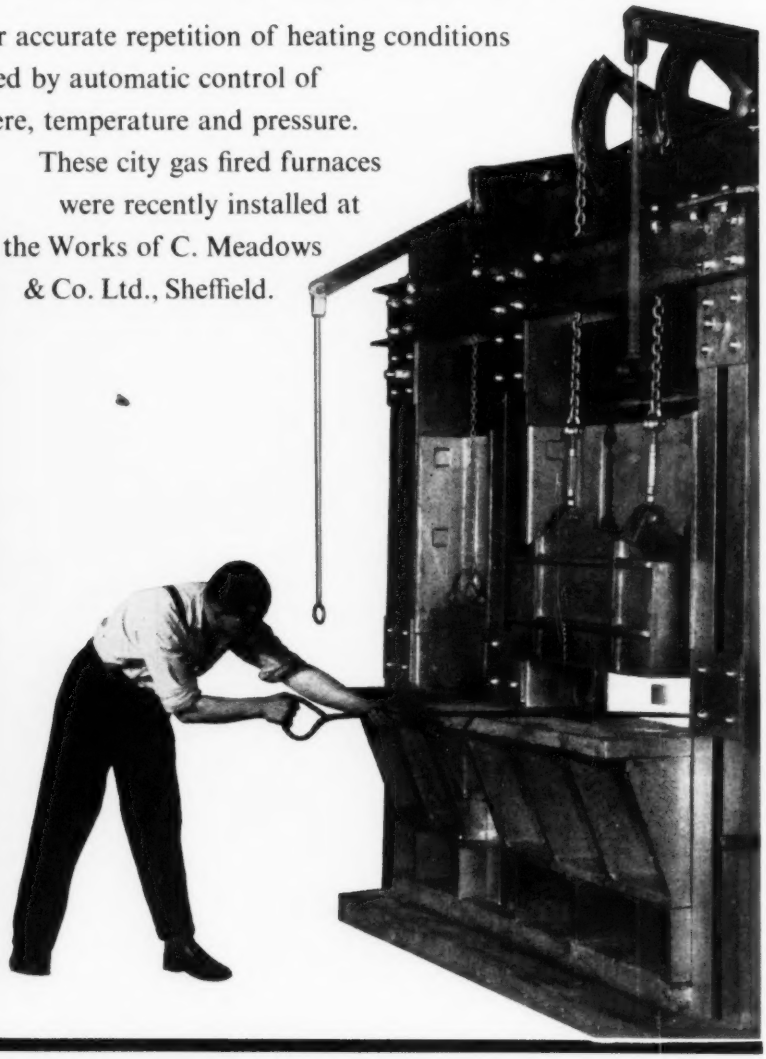

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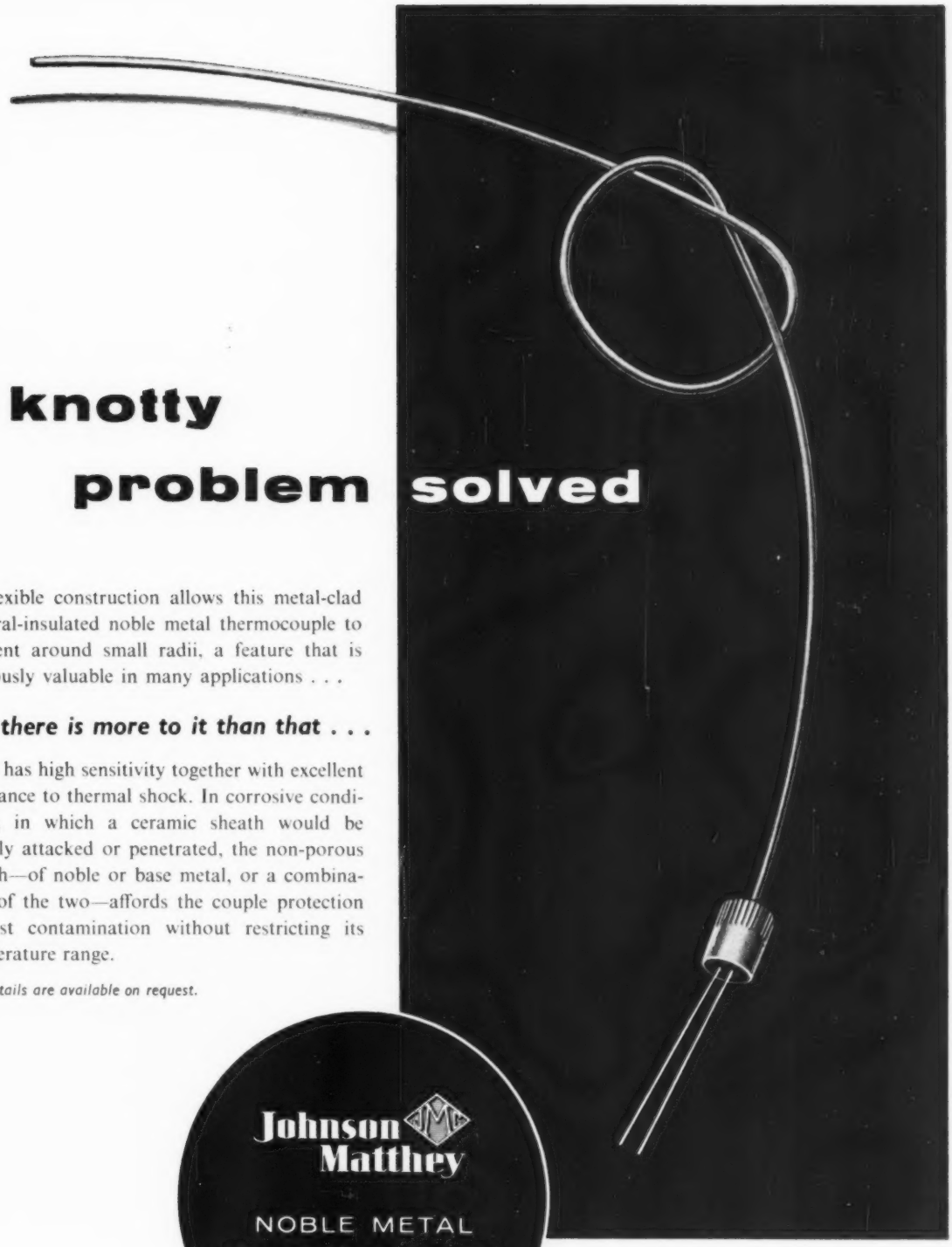

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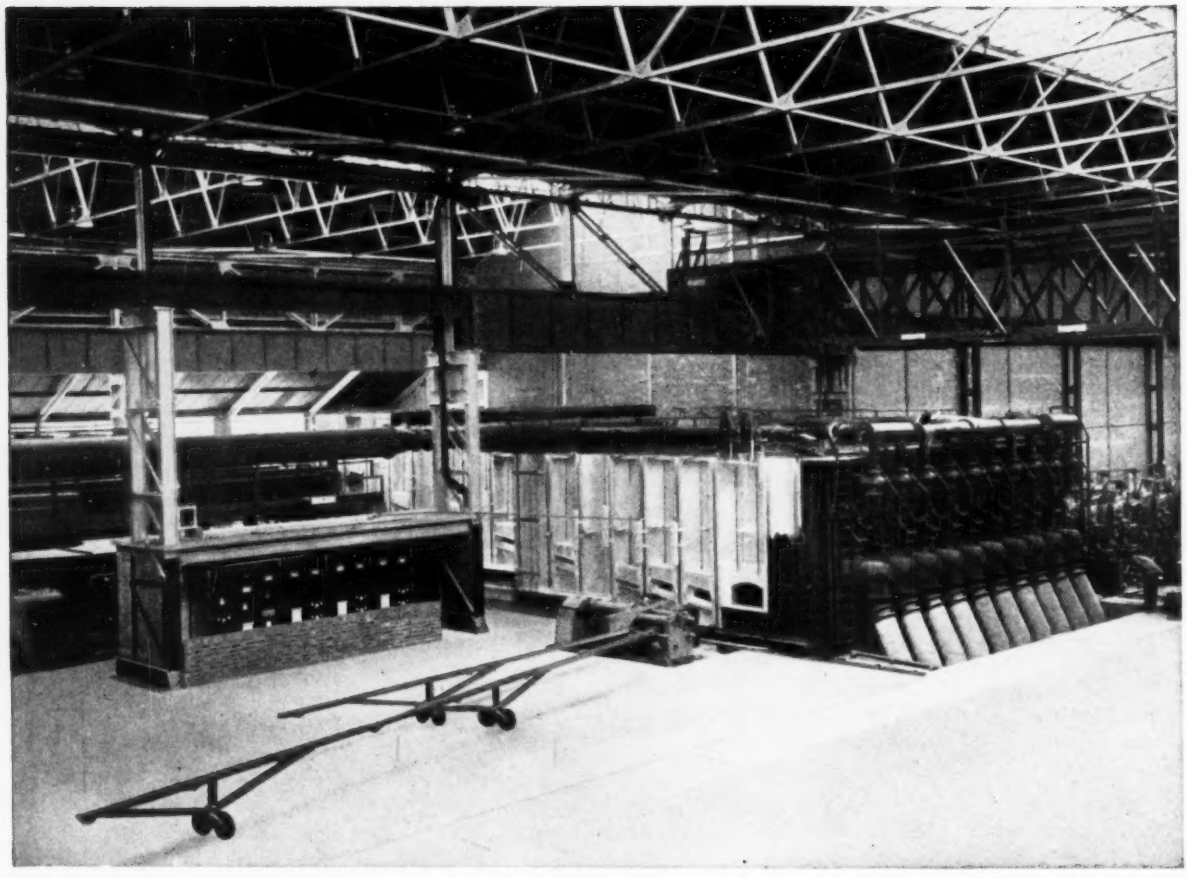

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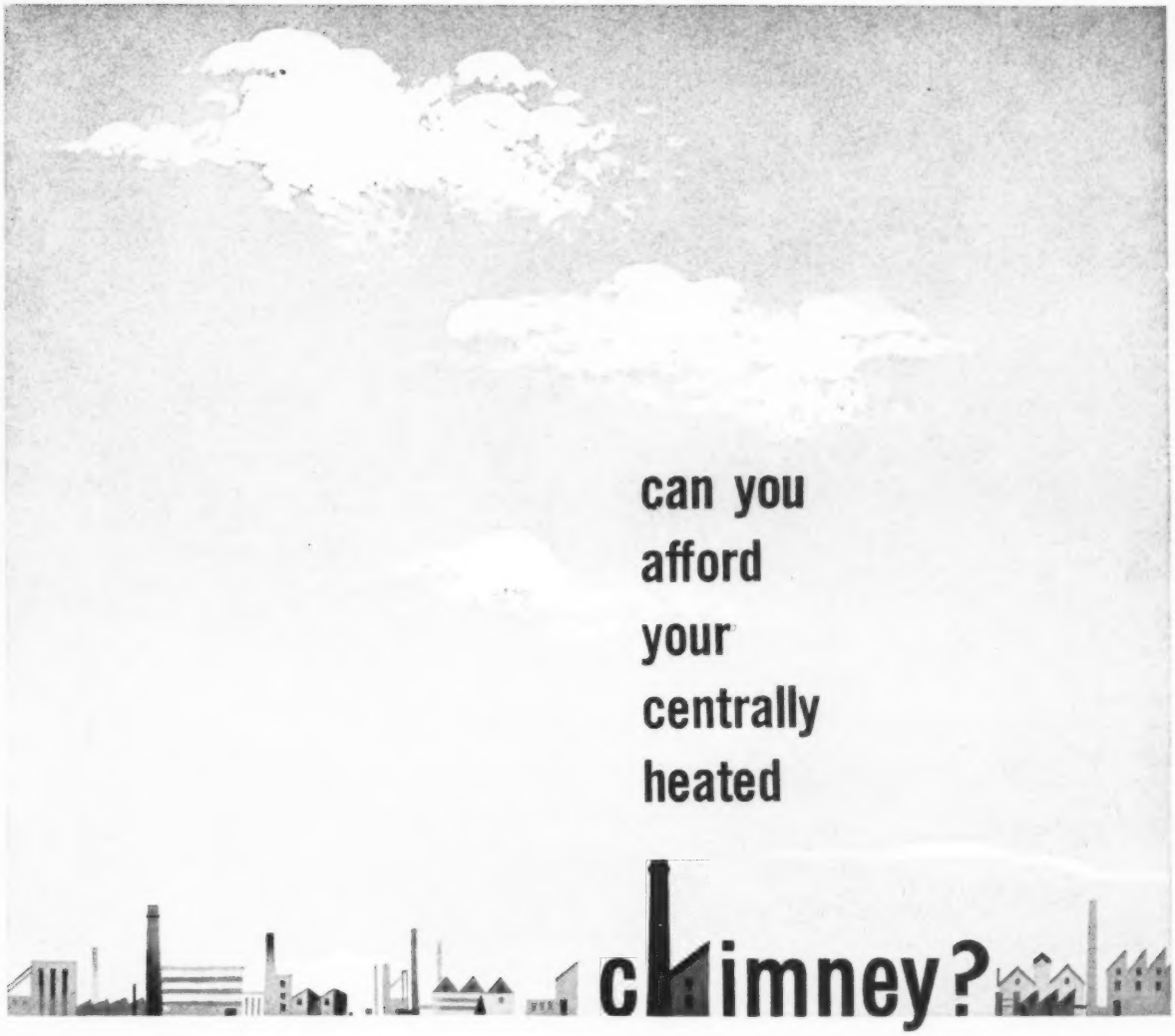

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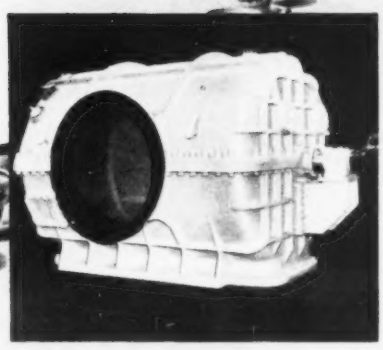

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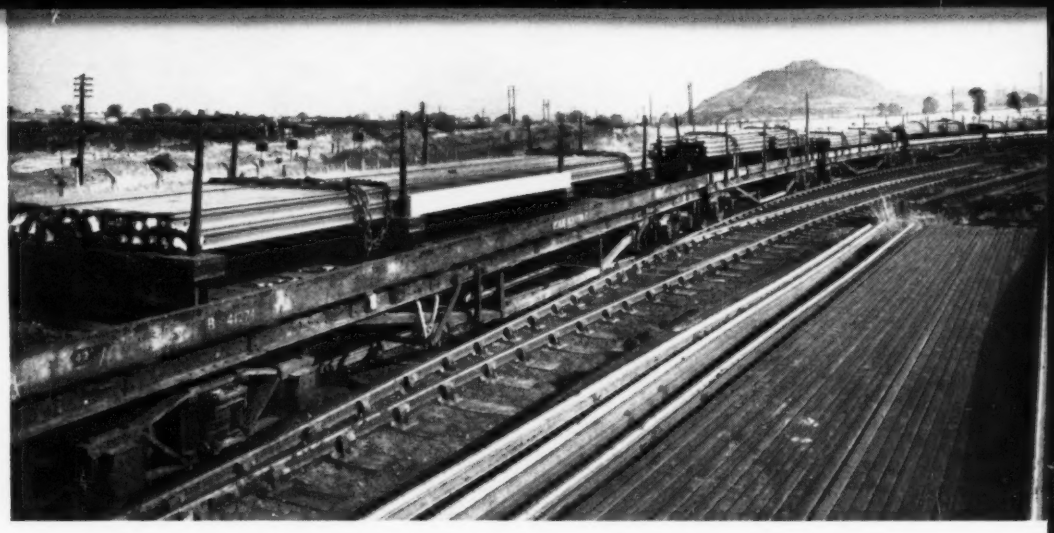

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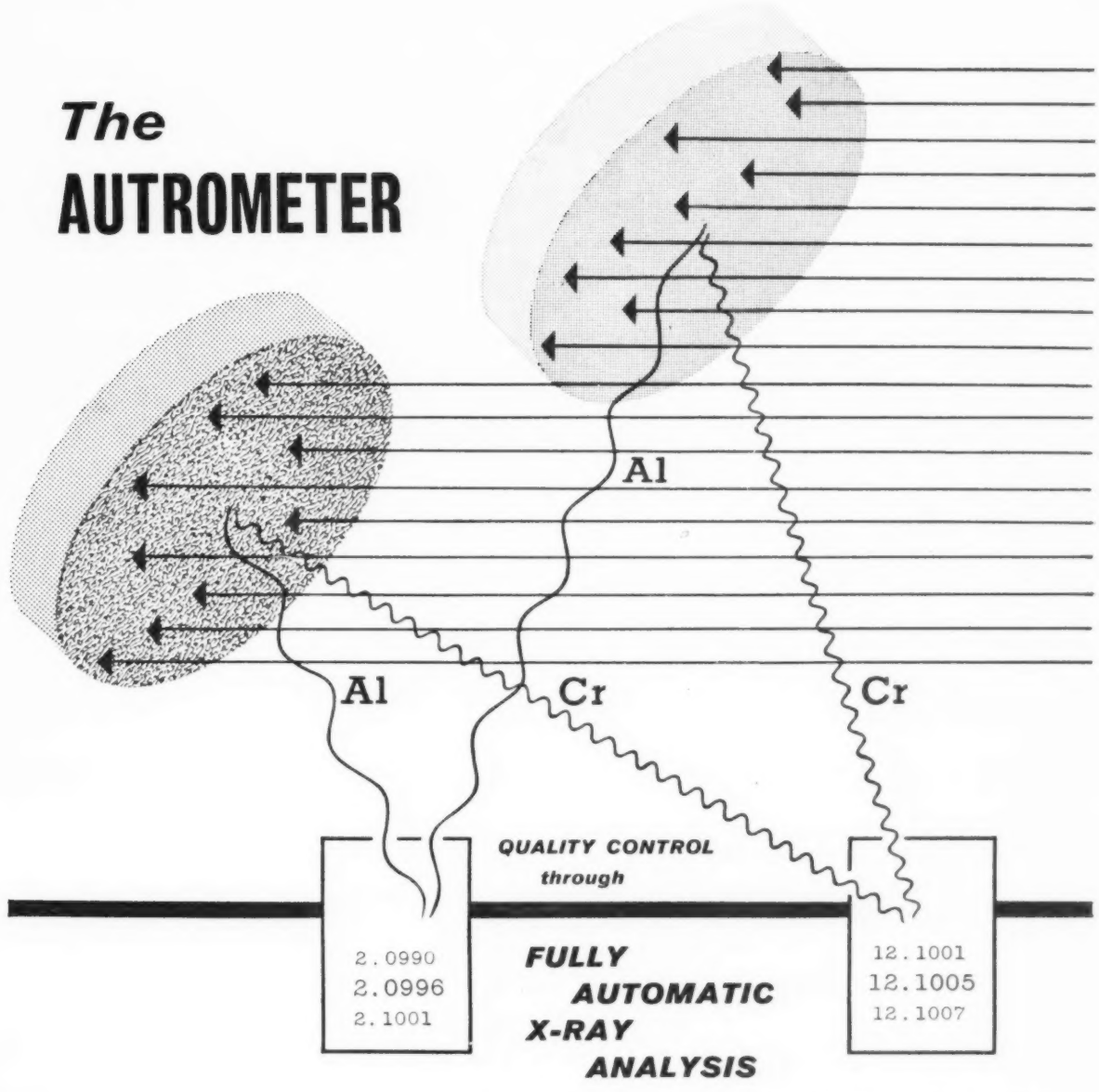

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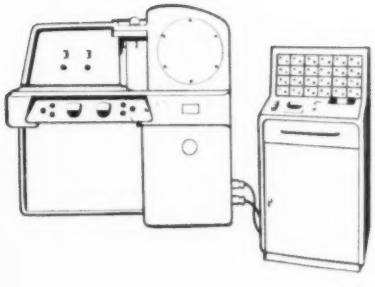

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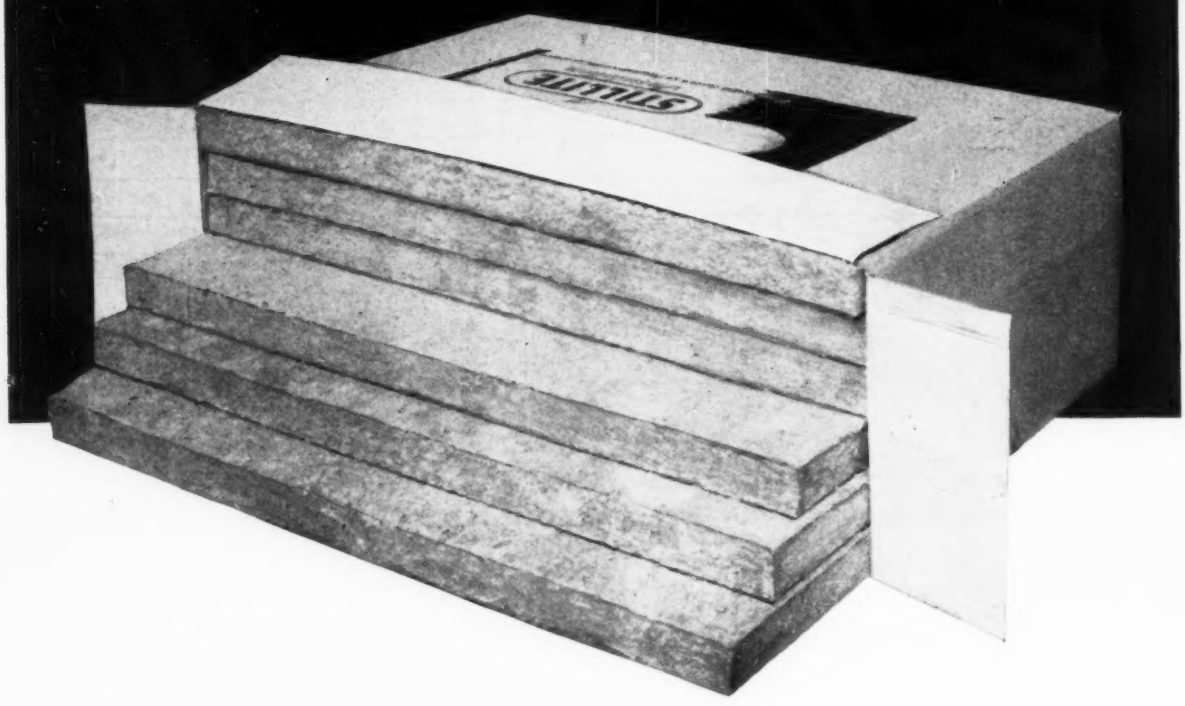

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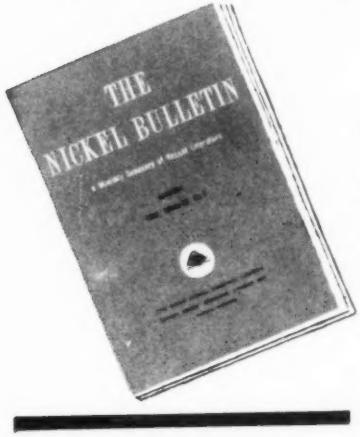

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New difficulties require new solutions. Even the very considerable know-how on steel amassed by the GKN Group is not *always* sufficient to solve our customers' problems immediately. That is why we—and our customers—attach such importance to our ability to find out when we do not already know how. Thus new products, new processes and new techniques to do with steel are the constant concern of all the GKN companies.

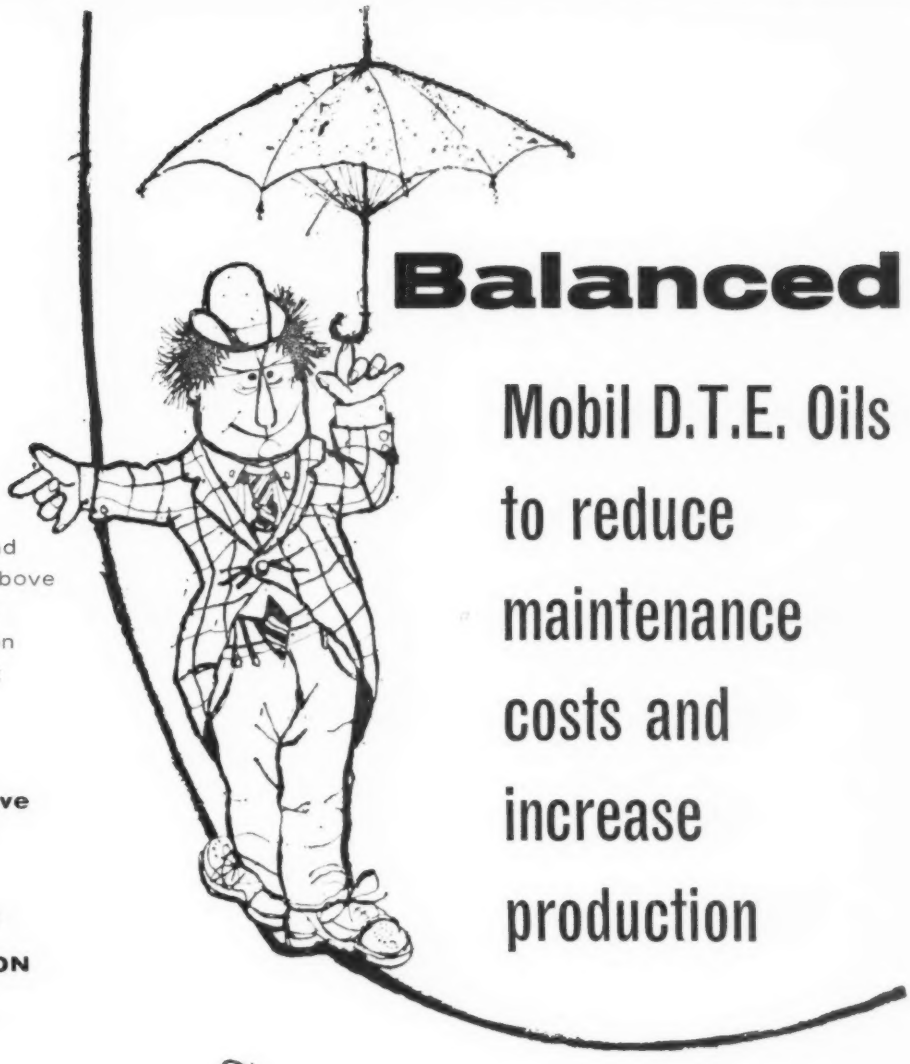
To strengthen the work of individual companies in this direction—and to help each company to maintain a lead in its own particular field—the Group Research Laboratory carries out research and development in mechanical and electrical engineering, metallurgy, physics, chemistry and instrumentation.

The Group Research Laboratory is excellently equipped and staffed. It helps to keep the Group up to date; and helps, too, to temper its less justifiable enthusiasms as well as its steel. It attracts new ideas wherever they are to be found. Behind each of the more than 80 companies in the GKN Group there is the Group Research Laboratory. That is one of the reasons why the Group can offer, with such confidence, anything in steel and steel products to practically anyone in the world.

THE GKN GROUP OF COMPANIES

FOR ANYTHING AND EVERYTHING IN STEEL

GUEST, KEEN & NETTLEFOLDS, LTD., HEAD OFFICE, LONDON WORKS, SMETHWICK 40, STAFFS. ENGLAND.



Balanced

Mobil D.T.E. Oils to reduce maintenance costs and increase production

Mobil D.T.E. Oils stand head and shoulders above all ordinary oils. They give *balanced* protection and performance that no oil on the market today can equal.

Mobil D.T.E. Oils give you the following advantages:

LONGER OIL LIFE

QUICK SEPARATION FROM WATER

FREEDOM FROM DEPOSITS

PROTECTION AGAINST RUST AND CORROSION

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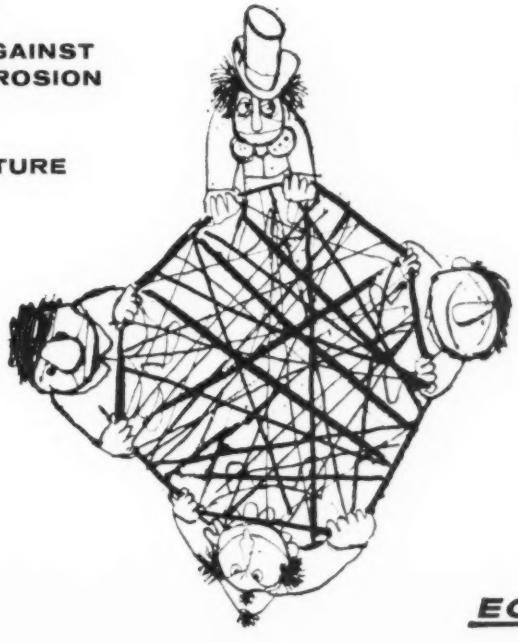


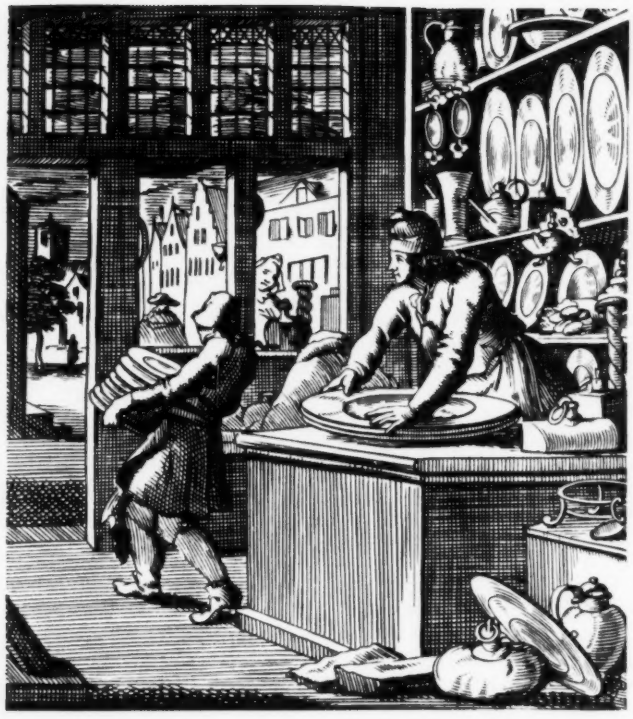
ASK THE MAN FROM MOBIL ABOUT BALANCED OILS

Our representative will give you the facts—the facts about the performance of Mobil D.T.E. Oils in comparison tests with other well-known hydraulic and circulating oils. Ask the man from Mobil to call and explain how Mobil D.T.E. oils can reduce maintenance costs and increase production

Mobil

ECONOMY SERVICE

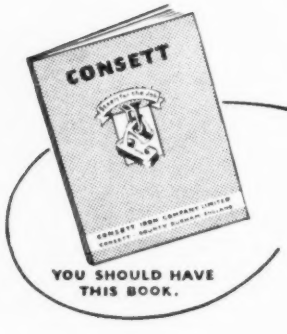
Reproduction of 16th century Dutch Print, illustrating the craft of the Tin-smith. (Copyright, Radio Times – Hulton Picture Library)



Every man to his own trade –

and our trade is STEEL!

Not only mild steel for everyday use but also the special steels demanded by the Nuclear Age . . . for example we produced the special steel plates for the reactor vessels at Calder Hall, Chapelcross and Bradwell, and have now supplied plates for the nuclear power station reactor at Latina, Italy.



An interesting booklet "Steels for the Job", just published by us, describes some of our newer products. We shall be delighted to send you a copy.

Our Technical and Research Department will also welcome enquiries from any whose problem is . . . **STEEL**

CONSETT IRON COMPANY LIMITED, CONSETT, COUNTY DURHAM

Telephone: Consett 341 (12 lines) Telegrams: Steel Phone Consett

The K.I.P. range includes lightweight insulating firebrick in 6 grades for temperatures up to 1500°C—'back-up' insulation up to 850°C and insulation castables up to 1300°C.

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ALL types of dust and fume control plant

Gases are cleaned and dusts controlled by Sturtevant plant to promote the vitally important interests of health, safety, economy and high production efficiency. The widest range of modern equipment is employed, including

ELECTROSTATIC PRECIPITATORS • CYCLONES

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Sturtevant plant is designed and installed to:

- CLEAN GASES AND COLLECT FINE DUSTS AND FUMES **from blast furnaces, open-hearth furnaces, sinter plant, converters, electric arc furnaces and hot blast cupolas**
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Enquiries to our reference Z/101/S

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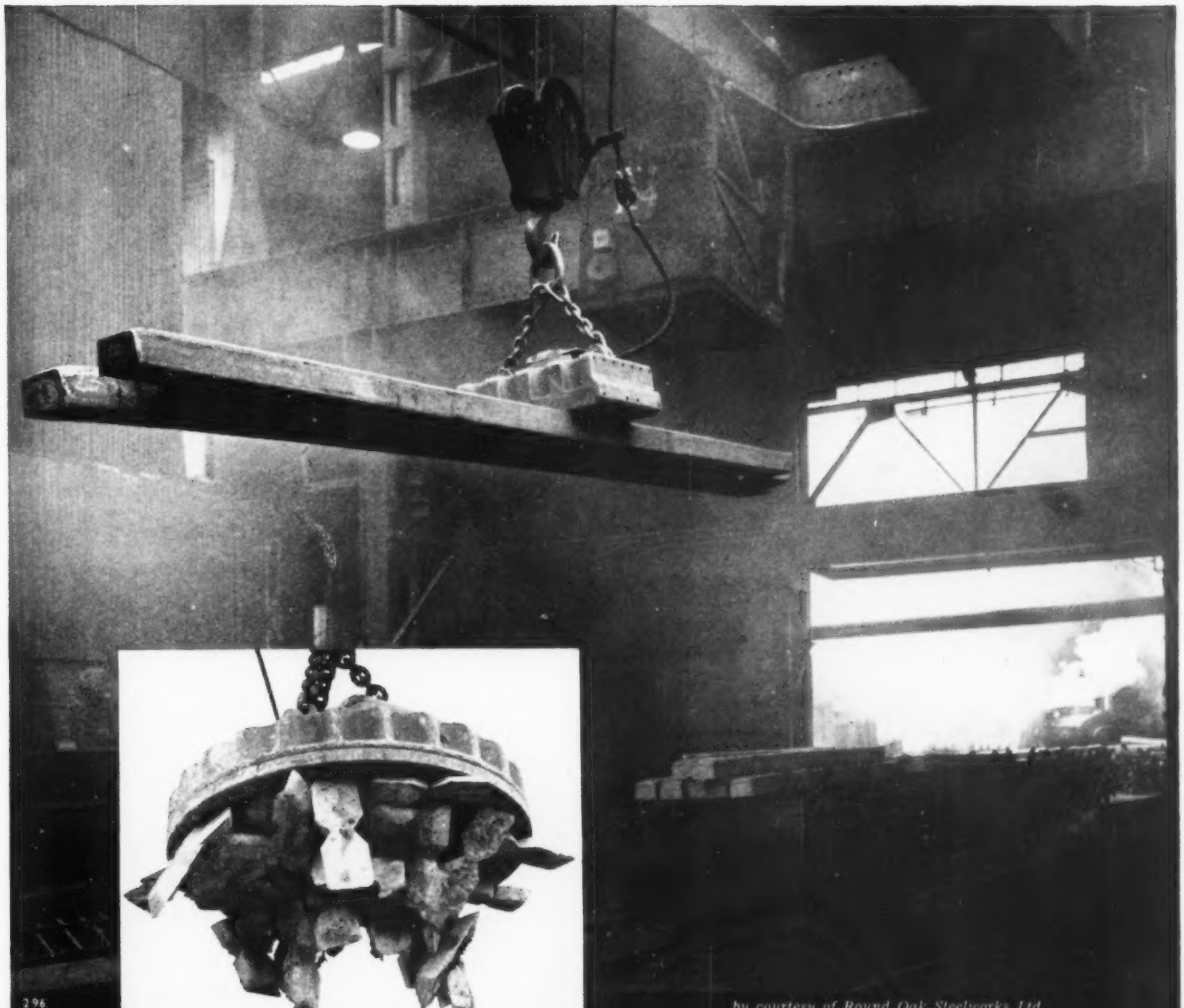
AUSTRALIA: STURTEVANT ENGINEERING CO. (AUSTRALASIA) LTD., 400 SUSSEX STREET, SYDNEY, N.S.W.



by courtesy of Round Oak Steelworks Ltd.

Circular, rectangular and bi-polar magnets are part of the wide range of electric magnets made by The General Electric Co. Ltd. Please ask for Technical Descriptions Nos. 315 & 379.

G.E.C.

Witton-Kramer lifting magnets

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When flue gases escape into the air they're gone for good, and with them goes the chance of a big fuel saving.

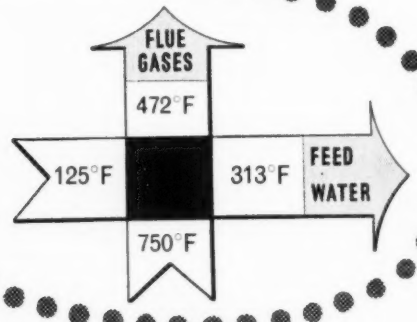
With a Senior Economiser you can use these gases to pre-heat the feed water, saving 10—20% of your fuel, increasing the efficiency of your boiler plant, cutting the cost of steam.

Senior Economisers, with their patented H-gilled heat exchange tubes, combine maximum effective heating surface with minimum resistance to flue gas flow. Straight, uniform gas passages, which do not collect soot, keep draught losses consistently low and rate of heat recovery constant.

If you have a boiler burning more than 20 tons weekly of any type of fuel, you can take advantage of Senior Economisers. They are built individually to suit your plant, and capital cost is soon repaid in fuel savings.

Why not write for further details?

Diagram shows temperature figures in a typical installation.



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for cheaper steam

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Northumberland House, 303-306 High Holborn, London, W.C.1
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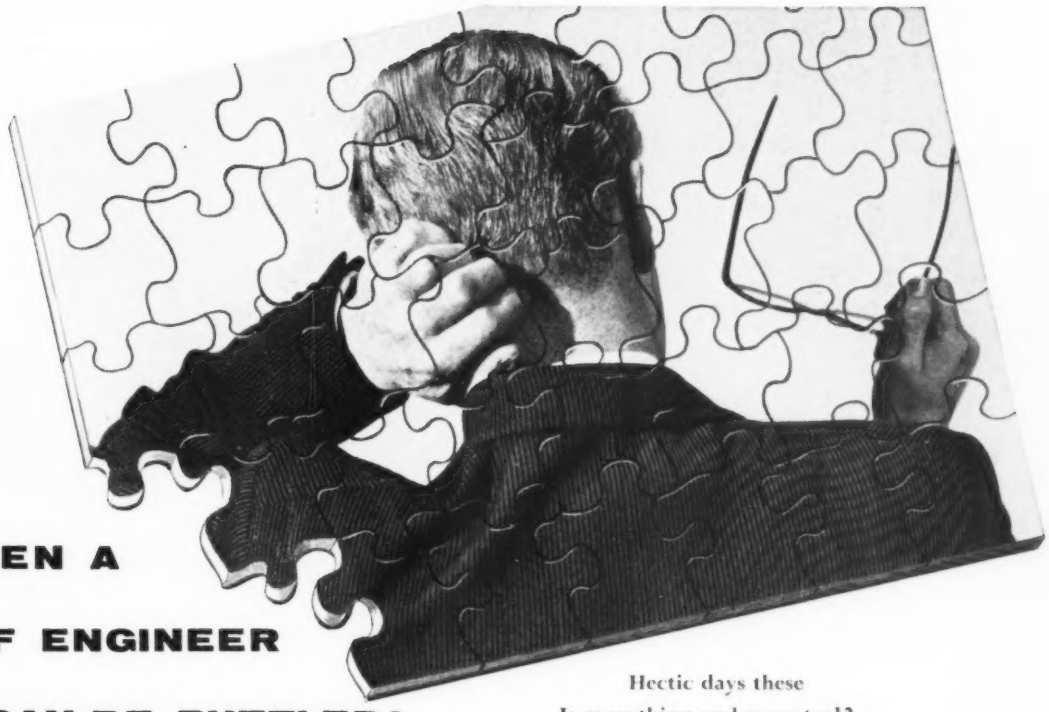

EVEN A CHIEF ENGINEER CAN BE PUZZLED!

Hectic days these
Is everything under control?
Pushbuttons, timers, isolators,
Electric motors in pumping room, casting room, plating shop
Mustn't fail or production loss, loss, loss
Processes integrated, sequences correlated
Is there a spanner in the house?
Electric, electronic, esoteric functions all,
Cosset them, cuss at them, keep them in hand
Motor selection, overload protection — don't forget
It's the wife's birthday tomorrow
J. B.'s new ideas for the finishing line; humour him,
Might be possible, though.
Mains wiring, control wiring, where's the labour coming from?
Control gear — DUPAR — that's the name
Must write.

SHED THE LOAD *of your motor control problems by consulting the Dewhurst Technical Advisory Service. There is DUPAR control equipment for all engineering and allied fields.*

DUPAR *for dependability*

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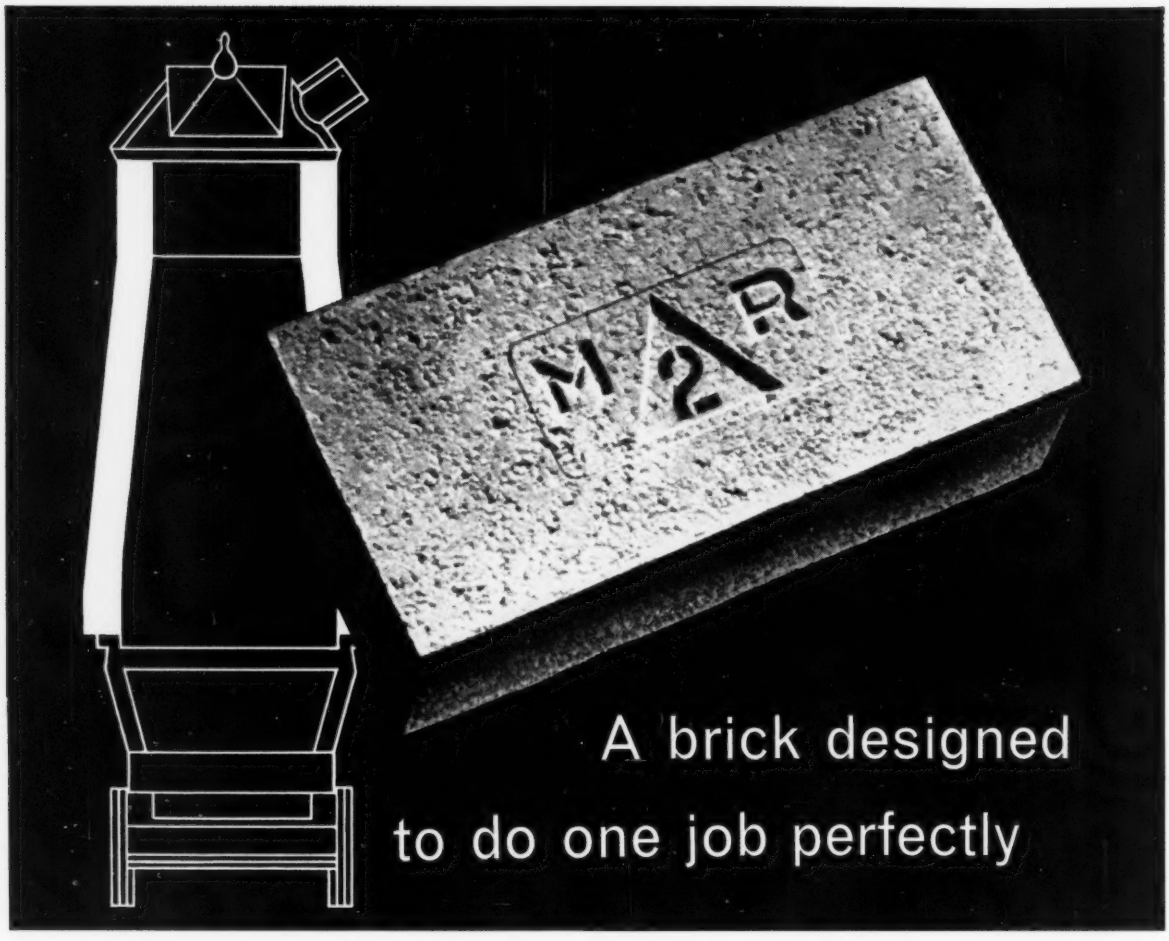
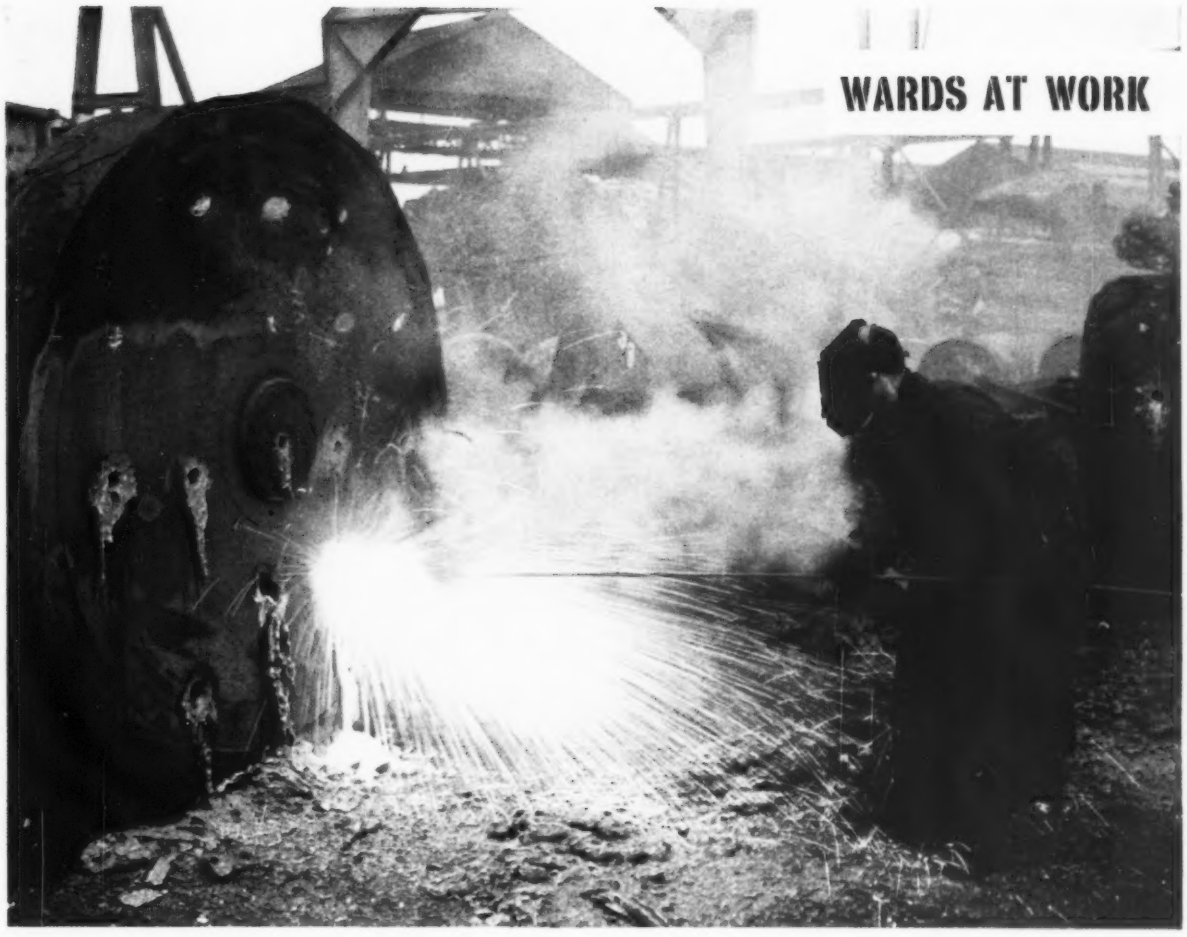


This is the M.R.2 Blast Furnace refractory – a very special brick indeed. It was developed for one application – the lining of Blast Furnaces operating on lean ores – with 100% sinter – and driven hard. M.R.2 is an extremely hard, low porosity refractory to enable it to resist the abrasive effect of the sinter. It is completely unaffected by carbon monoxide and is manufactured from very carefully selected raw materials prepared and moulded under the most stringent control with a final kilning at temperatures of the order of 1600°C to ensure complete volume stability at normal Blast Furnace operating temperatures. This is a brick with a narrow market, but it is typical of the way the Chemists, Physicists and Engineers at Neston look upon the making of refractories – as a branch of precise chemical engineering. Amongst the other things they make with equal care are M R.1 and M.R.60 bricks and special shapes; low-heat storage insulating refractories M.I.28, and the whole range of Tri-Mor Mouldables and Castables for Monolithic linings which are having such a profound effect on all classes of furnace design.

MORGAN Refractories Ltd

MORGAN REFRACTORIES LTD · NESTON WIRRAL CHESHIRE · TELEPHONE: NESTON 1406



He turns base metal into gold!

No alchemist this, but a skilled scrapman in a Ward scrapyard. With his highly specialised experience, he can quickly prepare large masses of scrap for reduction to furnace sizes. Not only steel and iron, but all non-ferrous metals are handled too, both under strict metallurgical control. At scrapyards all over the country, equipped with the most modern plant, Wards handle thousands of tons of metal a day, and they are proud to be the leaders in this important industry.

Helping to feed the furnaces with carefully graded scrap is only one activity of the Ward Group of Companies whose products and services cover almost every branch of industry.

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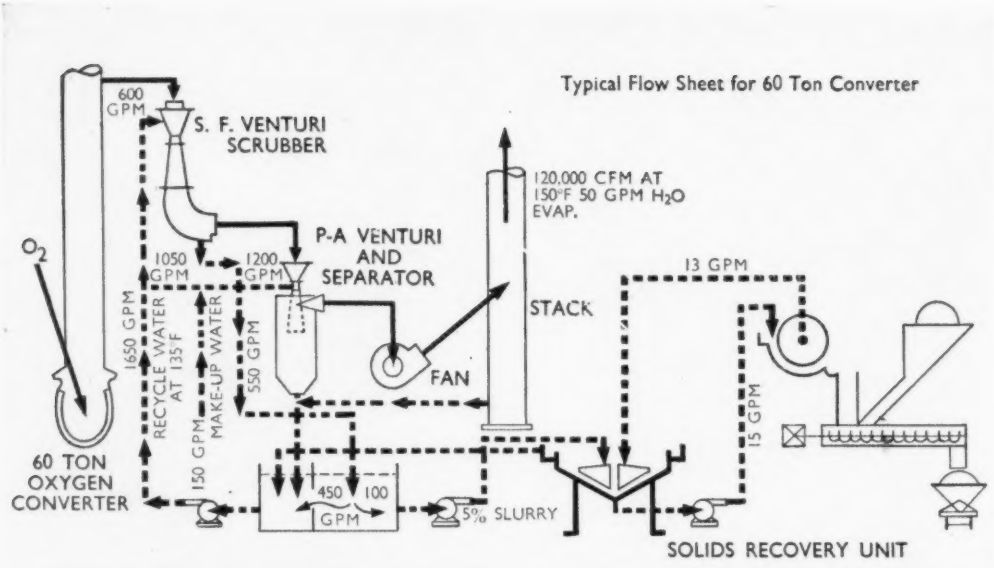


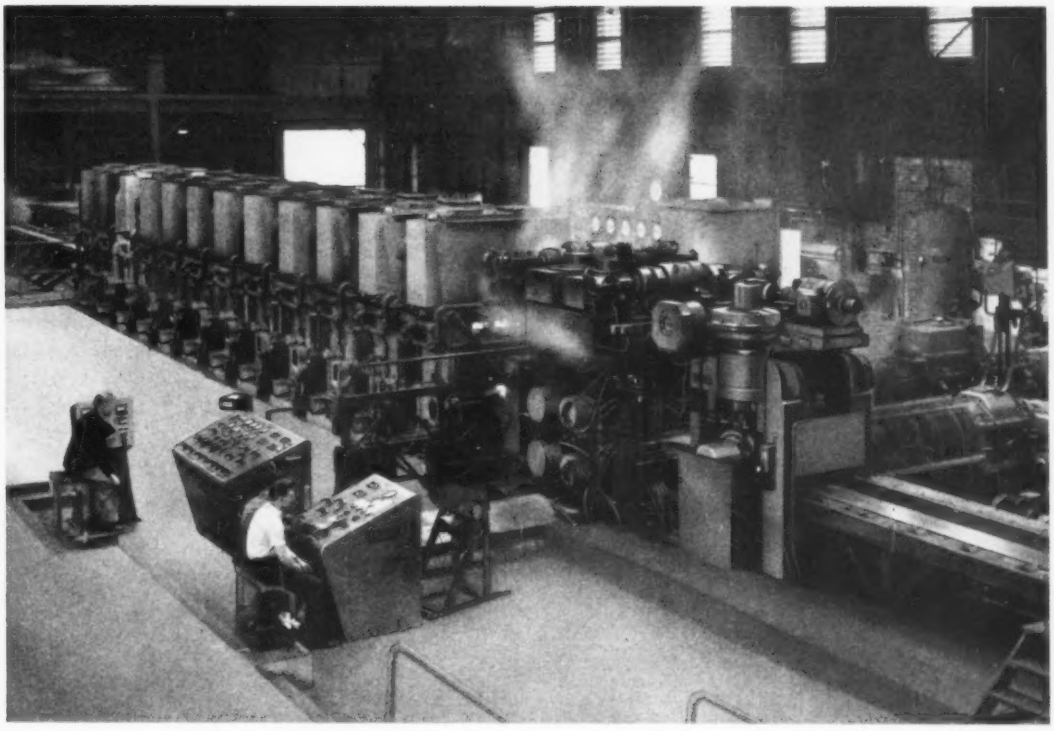
six converter installations prove

The recent installation of Chemico Venturi Gas Scrubbers at a major steel plant in Canada marks the sixth time Venturi Gas Scrubbers have been used to prevent an extremely undesirable source of air pollution—Iron Oxide Dust. Chemico Venturi Scrubbers maintain leadership in the world's steel industries because they offer wet scrubbing which allows non-explosive collection, efficiencies above 99% and extremely low maintenance. In addition no critical controls are required to maintain efficiency and the handling of dust is simplified by recovery in a dry form. Chemico Gas Scrubbers are ideally suited for Open Hearth Furnaces, Cupolas, Scarfing operations, Sintering operations, and Electric Furnaces. For further information and a copy of a brochure giving full particulars, contact the Chemico Gas Scrubber Division:

CHEMICO

CHEMICAL CONSTRUCTION (G.B.) LTD. 9 HENRIETTA PLACE, LONDON, W.1. LANGHAM 6571

A SUBSIDIARY OF CHEMICAL CONSTRUCTION CORPORATION, NEW YORK, U.S.A.



SENDZIMIR PLANETARY HOT ROLLING MILL

BUILT BY **ROBERTSON**

FOR MESSRS. J. J. HABERSHON & SONS LTD., ROTHERHAM
BY WHOSE COURTESY THIS PHOTOGRAPH APPEARS

This mill is designed to produce hot rolled strip up to 15 in. wide and as thin as ·048 from slabs 1¾ in. thick in mild, high carbon and stainless steels.

The mill consists essentially of feed rolls, planetary roll assembly and a final two-high planishing mill followed by coiling equipment.

Nominal output is 10 tons per hour which has already been considerably exceeded.

Surface finish is fully up to the standard required for subsequent cold rolling.

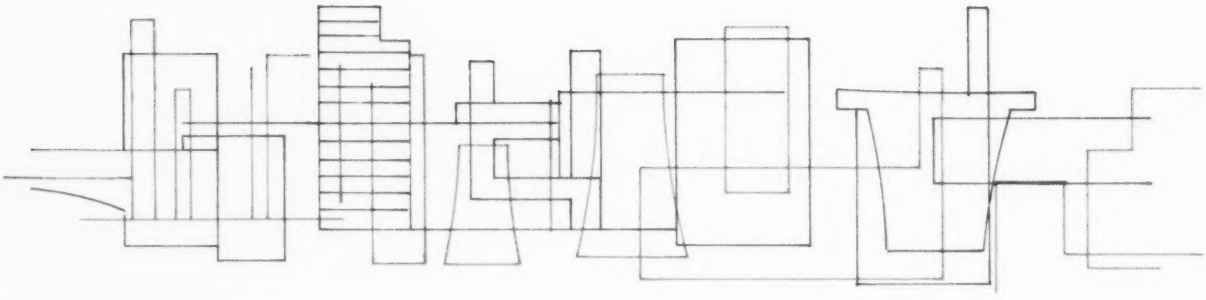
This is the second Planetary Mill in Great Britain both of which have been built by Robertson.

W. H. A. ROBERTSON & CO. LTD., Bedford, England

LICENSEES FOR THE BUILDING OF SENDZIMER COLD REDUCTION MILLS AND PLANETARY HOT MILLS, HALLDEN GUILLOTINE AND ROTARY FLYING SHEAR MACHINES AND TORRINGTON METAL WORKING MACHINERY.

WT. 311R

The Colville Group of Companies



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In Scotland, Colvilles are engaged on a vast development programme. By 1963 ingot capacity will be raised to 3.3 million tons per annum, and the Group will produce in their modern mills the full range of steel products—heavy and light plates, sheets and strip. Heavy and light sections, bars and broad flange beams.

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SOMETHING NEW IN SHEET STEEL . . . ready surfaced with durable vinyl in the colour of your choice, all set for fabrication—that's **Stelvetite.** It can be cut, deep drawn, folded flat and welded; and the hard-wearing vinyl face which **Stelvetite** shows the world will resist oils, detergents, acids—and time. It comes in a wide range of colours and textures. It is made into vehicles, partitions, ceilings, furniture, kitchen equipment, wall cladding and drums . . . all of them colourful and durable. In production costs alone **Stelvetite** is a money saver. Investigate it!

Stelvetite

Plastic bonded to Steel

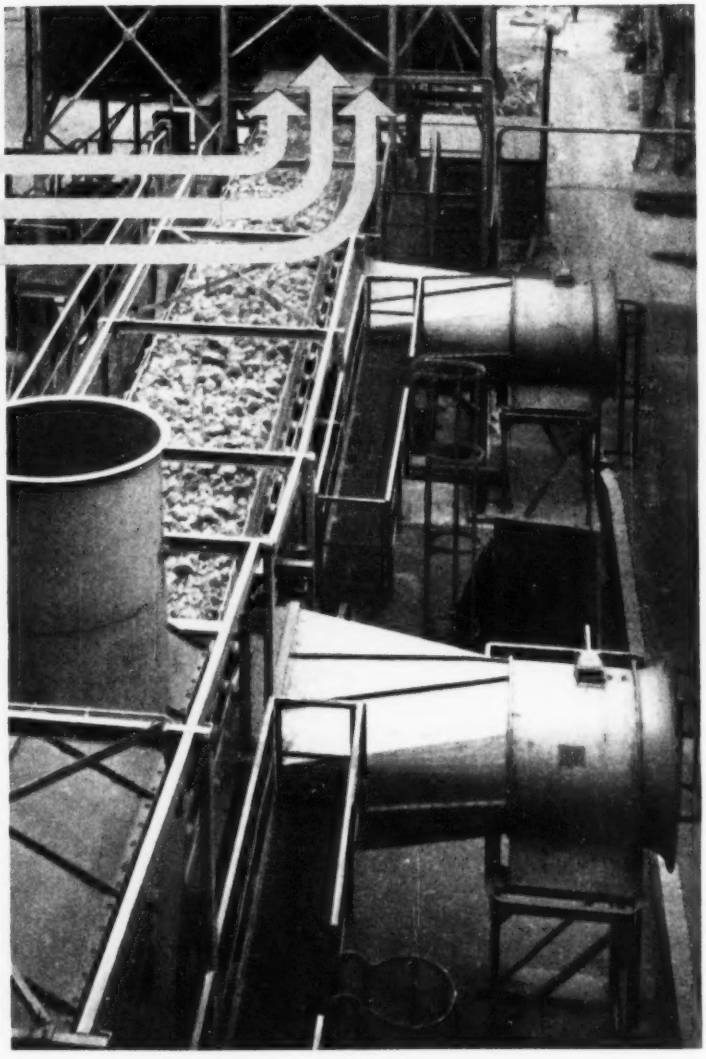
* *Accepted by the Council of Industrial Design for "Design Index."*

All enquiries about Prestcold refrigeration units should be made to: Prestcold Division, Pressed Steel Company Ltd., Swansea, Glam.

STELVETITE*—made in co-operation with BX Plastics Ltd—**by John Summers & Sons Ltd.** who produce over 52,000 miles of sheet steel every year. Write to us at Dept. JS, Shotton, Chester, for full information.

SPEEDING THE SINTER FLOW

This photograph was taken looking down on the sinter cooler at Barrow Ironworks. On the right of the cooler can be seen two of the three fans which have reduced the cooling interval so efficiently

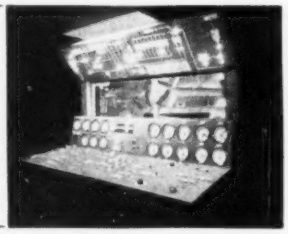


For 70 years we have made steady improvement in the sintering process our business, and each sinter plant we design and install incorporates the latest developments in sintering technique.

At many new sinter plants, including those at Barrow Ironworks, Consett Ironworks, the Iron and Steel Corporation of South Africa, Cockerill-Ougrée (Belgium) and the Société Métallurgique de Normandie (France), a new Huntington Heberlein cooler has increased the efficiency of the sinter process. High quality sinter is being cooled in 20 minutes from 600°C for transmission on normal belt conveyors. This cooling is achieved by forced circulation from cold air fans which work at very low power consumptions.

Associated in the U.S.A. with Koppers Company Incorporated, our other specialities include: Ore Dressing Plants, High Intensity Magnetic Separators, Herreshoff Furnaces and Badische Turbulent Layer Roasters.

The feed rate control panel at Barrow



CONTINUOUS SINTERING PLANTS BY

Huntington, Heberlein & Co. Ltd

Simon House, 28-29 Dover Street, London, W.1. Telephone: Hyde Park 8191. Telex: 2-3165. Telegrams: Innovation Wesphone London Telex

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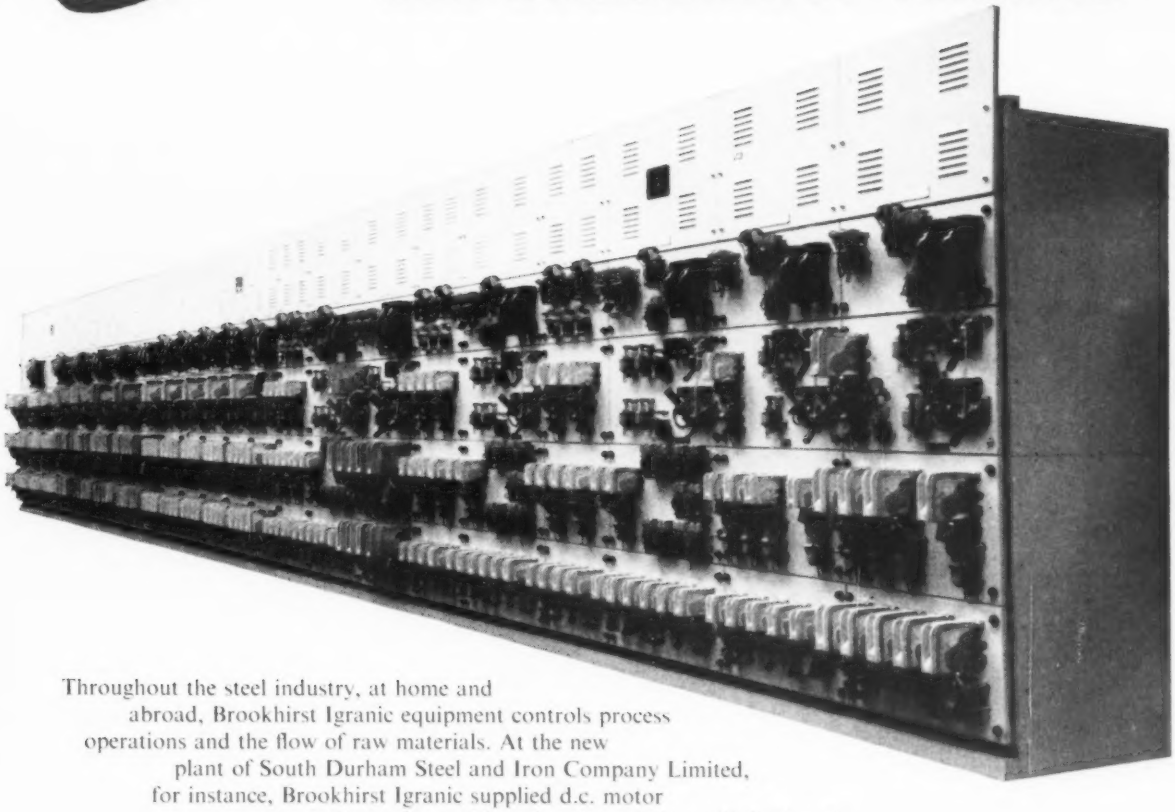
HH29

Steel...

and BROOKHIRST IGRANIC



Throughout the steel industry, at home and abroad, Brookhirst Igranic equipment controls process operations and the flow of raw materials. At the new plant of South Durham Steel and Iron Company Limited, for instance, Brookhirst Igranic supplied d.c. motor control gear for soaking pits, the slabbing mill, light and heavy shearing lines, and d.c. and a.c. electrical charging control gear for the blast furnace.

There can be no finer tribute than this to the quality and reliability built into Brookhirst Igranic control.

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Metal Industries Group

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Makers of Britain's widest range of electrical control and associated equipment

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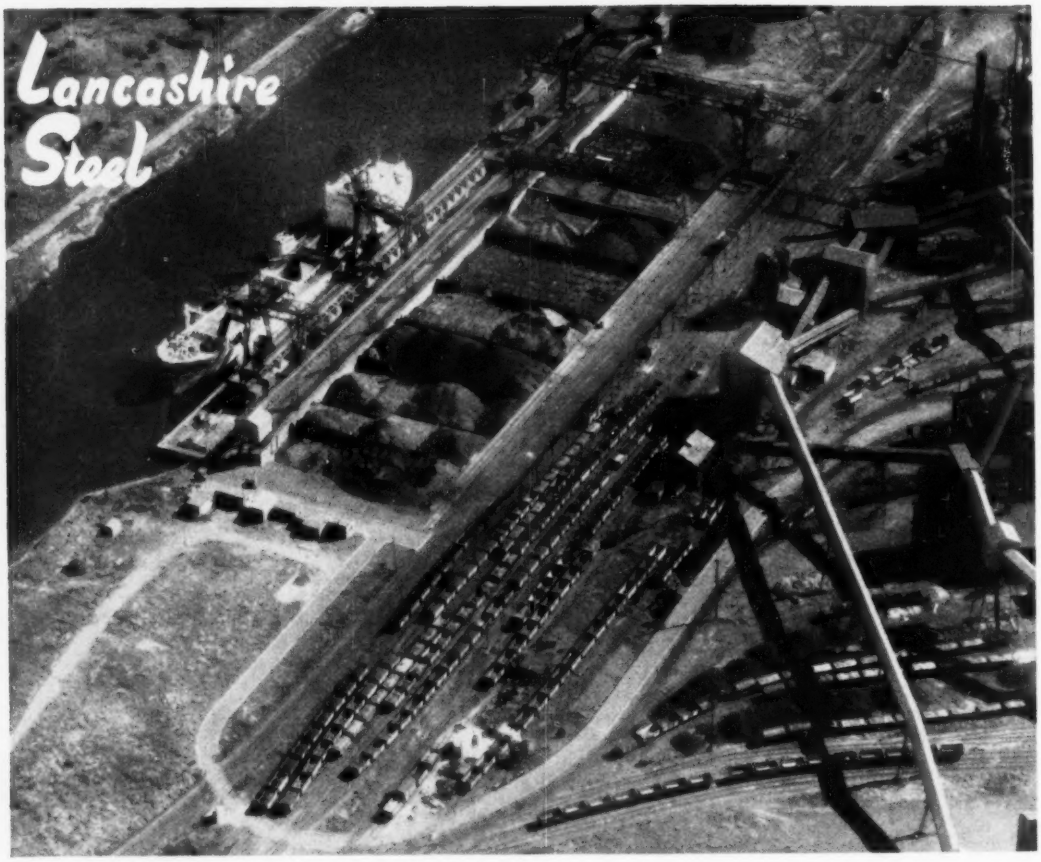

AERIAL VIEW OF THE UNLOADING WHARF—*Photograph by Aero Pictorial Ltd.*

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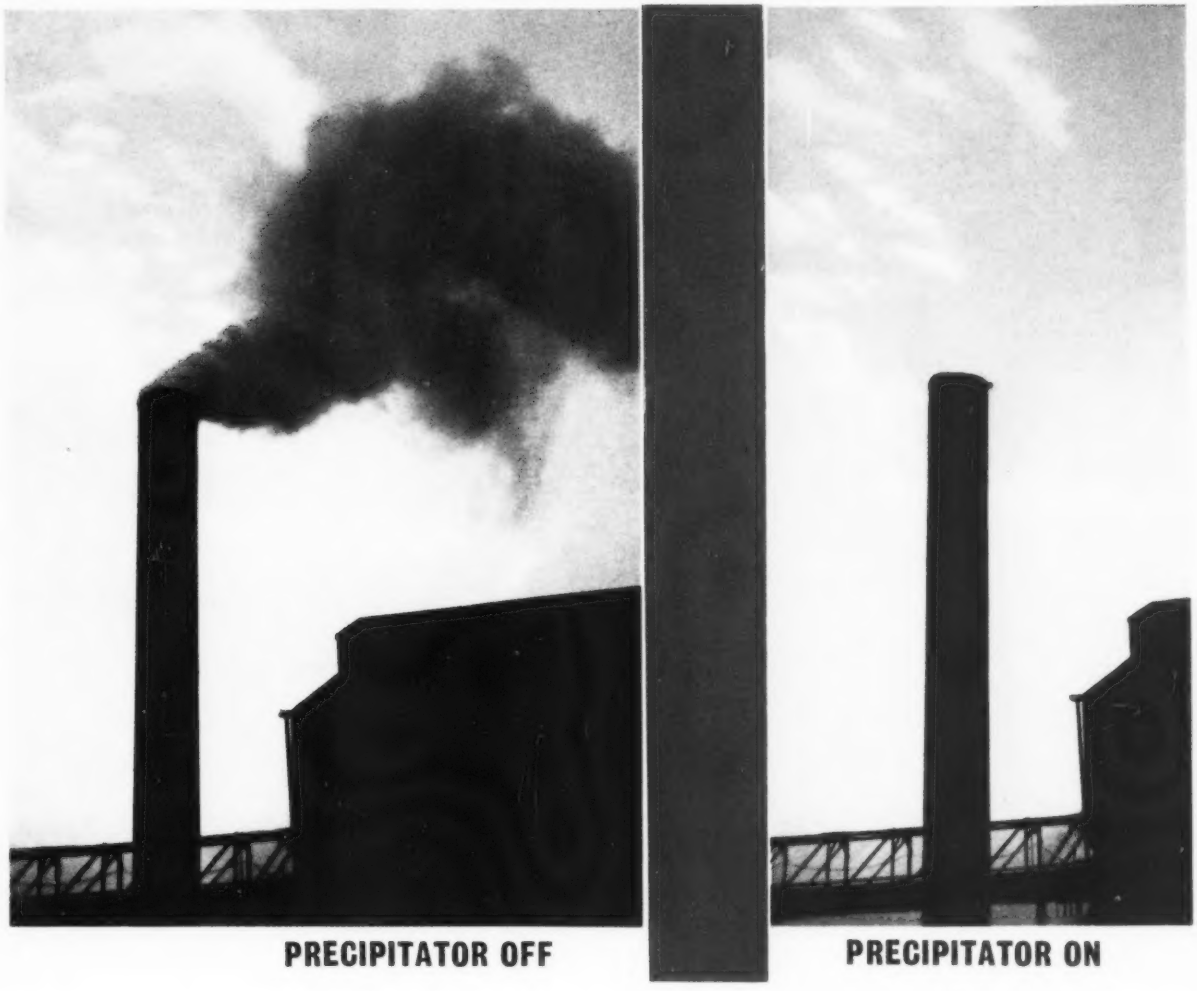
IRON OXIDE FUME REMOVAL FROM DESEAMING PLANT



PRECIPITATOR OFF **PRECIPITATOR ON**

The photographs reproduced above were each taken at the moment of peak fume production from the scarfing process on two successive cycles of operation. They give a clear indication of the efficiency of the Holmes-Elex Electrical Precipitator which is cleaning exhaust air from the deseaming plant at the Rotherham Works of Messrs. Steel Peech & Tozer (Branch of the United Steel Companies Limited).

The first electrical precipitator to operate on deseaming plant in the United Kingdom was installed by our Gas Cleaning Division at the Corby Works of Messrs. Stewarts & Lloyds Limited in 1954. A further installation was recently commissioned at the Bilston Works of the same company.

Reproduced by permission of Messrs. Steel Peech & Tozer.

W. C. HOLMES & CO. LTD.

Gas Cleaning Division, Turnbridge, Huddersfield

Telephones: Huddersfield 5280
London: Victoria 9971
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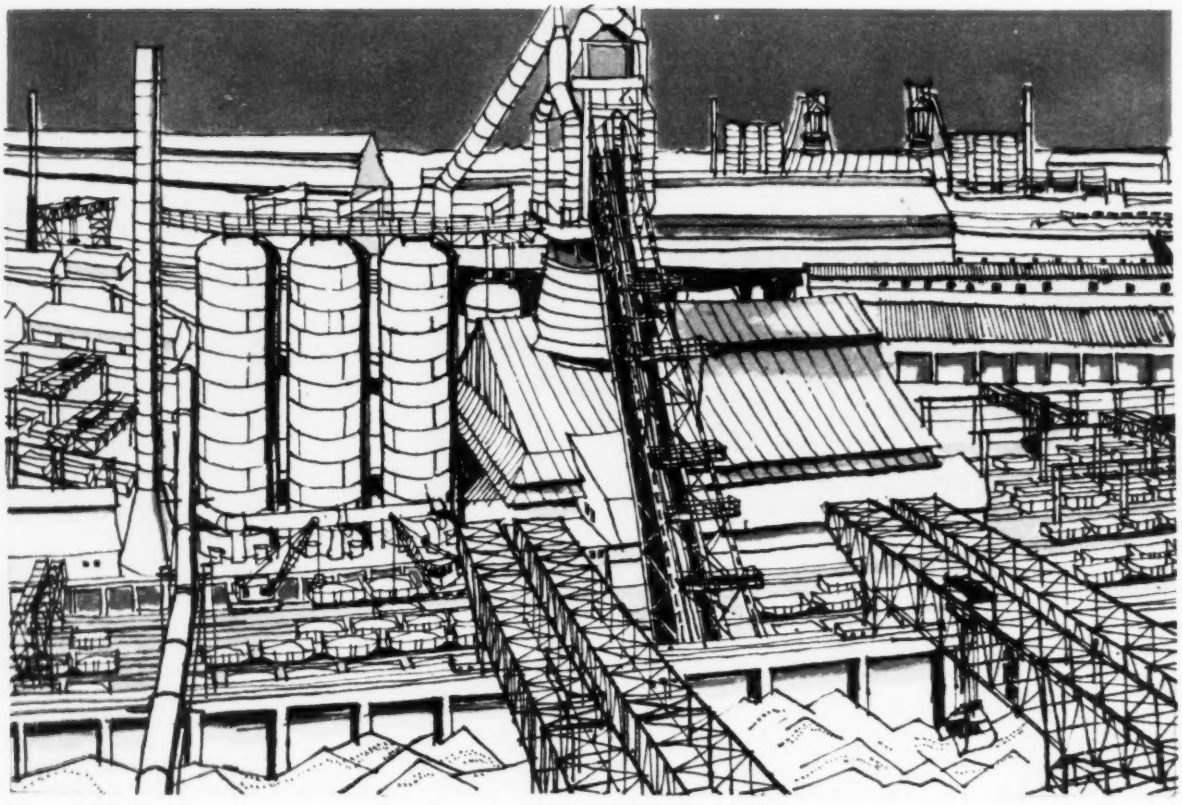

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and general purpose
steelworks handling equipment

General Purpose overhead
travelling cranes to Steelworks
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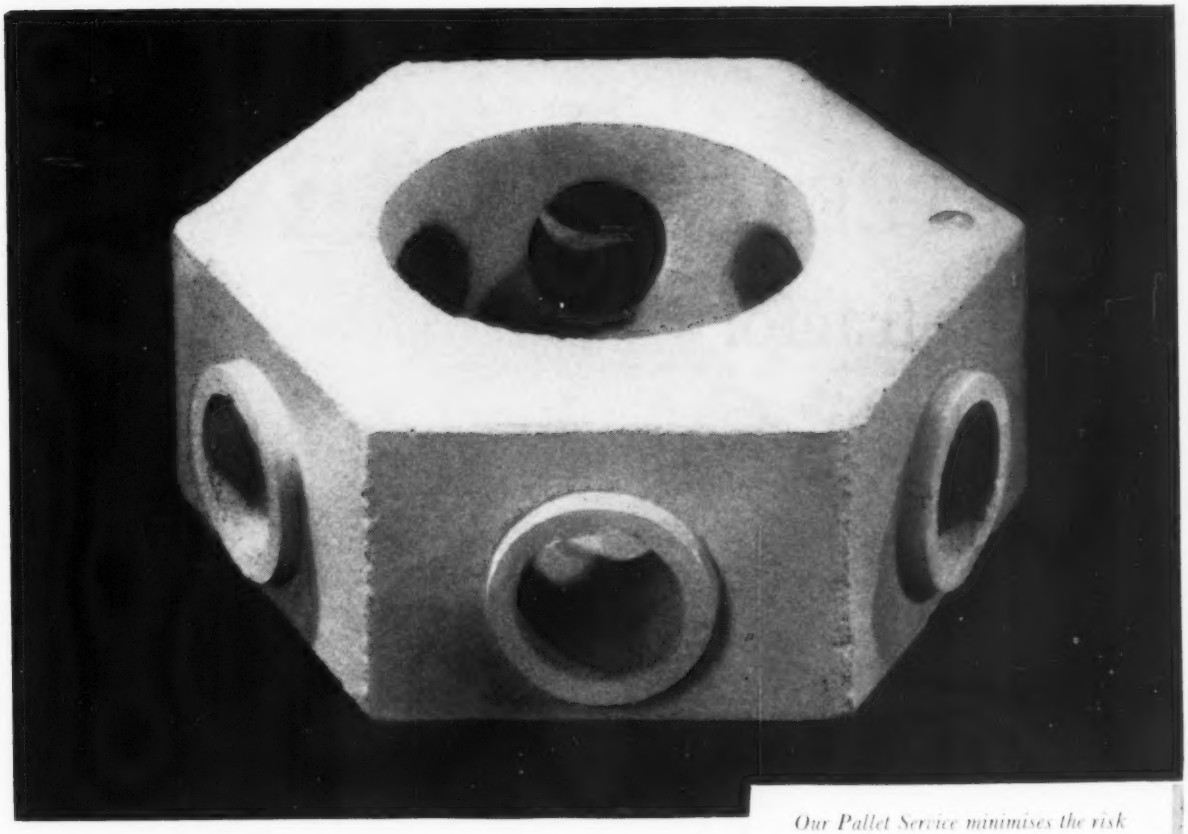

Centre of attraction

To the expert's discerning eye, a centre brick made by Wraggs of Sheffield represents all that's finest in casting pit refractory manufacture. In addition to perfect external shape and accuracy of dimensions, the smooth internal finish ensures that no refractory inclusions are picked up by the metal during pouring.

These attributes stem from years of specialisation in the casting pit refractory field, in which Wraggs were pioneers. And they all add up to the solid reliability demanded by to-day's busiest steelmakers.

Our Pallet Service minimises the risk of damage in transit — saves time and trouble in handling. Particulars gladly sent on request.

Wraggs of Sheffield

for reliable casting pit refractories

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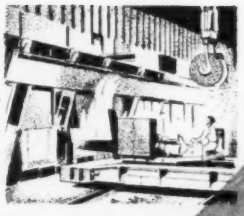

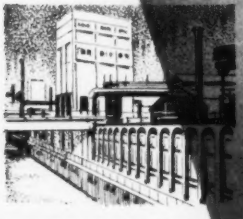

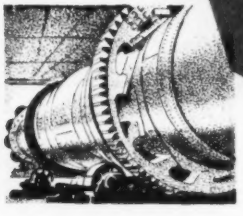

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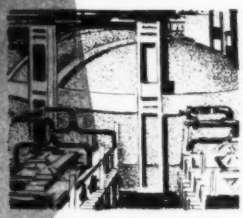

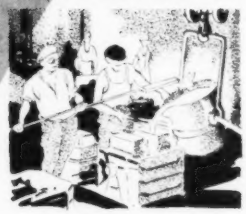

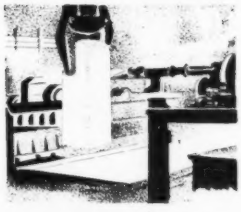

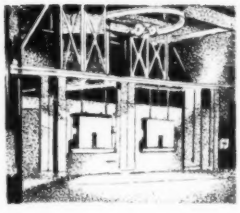

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Contents

THE IRON AND STEEL INSTITUTE

Creep, mechanical, and metallurgical properties of an experimental Mo–V steel rotor forging 117
A. I. Smith, E. A. Jenkinson, D. J. Armstrong, and L. M. T. Hopkin

The electrolytic production of sulphur-free iron 131
W. T. L. Neal, J. F. Duke, and H. R. Sullivan

Factors affecting the adherence of oxides on metals 135
R. F. Tylecote

The kinetics of the martensite transformation in iron alloys 142
W. S. Owen and A. Gilbert

The mechanical properties of some tempered alloy martensites 149
J. M. Capus and G. Mayer

Dislocations in α-iron 160
D. G. Brandon and J. Nutting

CORRESPONDENCE ON PAPERS

The creep and rupture properties of $2\frac{1}{4}$% chromium–1% molybdenum quality steel 158
J. D. Murray, J. S. Blair, G. G. Foster, H. W. Kirkby, and J. Blackhurst

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

The effect of heat-treatment and microstructure on the high-temperature ductility of 18%Cr–12%Ni–1%Nb steels 166
K. J. Irvine, J. D. Murray, and F. B. Pickering

Some ductility aspects of 18–12–1Nb steel 180
R. J. Truman and H. W. Kirkby

Heat-affected zone cracking in welded high-temperature austenitic steels 188
R. N. Younger and R. G. Baker

IRON AND STEEL ENGINEERS GROUP

Investigation of failures in a blooming mill 195
F. E. Stacey

Some aspects of lighting in the iron and steel industry 207
D. R. M. Nisbet

Engineering aspects of the Redbourn development scheme 213
P. M. Hesketh

Developments at Appleby-Frodingham since 1954 217
E. A. Atkin and E. F. Farrington

News 225 **Translations** 228 **Abstracts** 229

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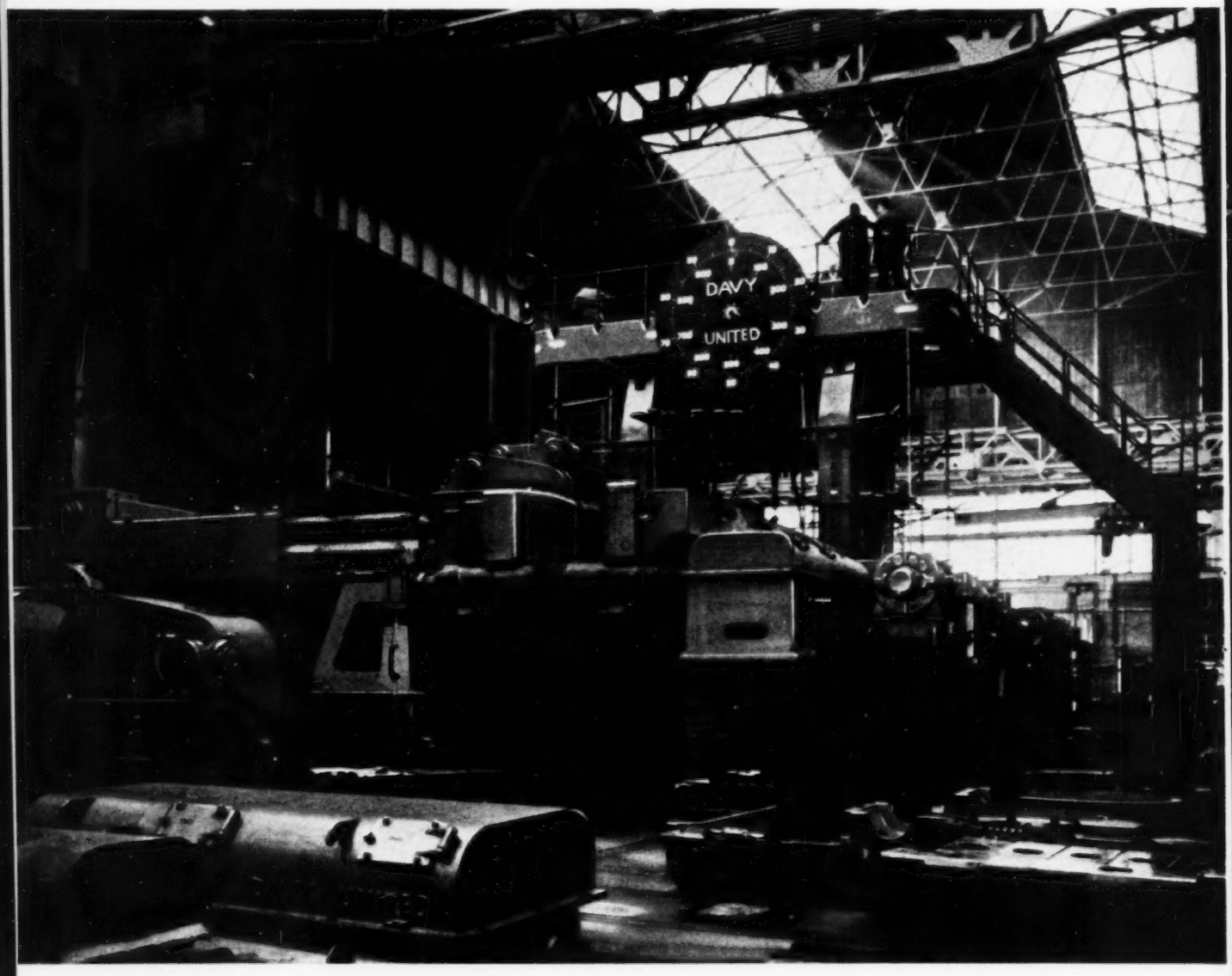
42" BLOOMING MILL FOR THE NEW STEELWORKS AT DURGAPUR

in West Bengal being constructed by ISCON, photographed in course of erection in Davy-United's Darnall Works. As a member of ISCON Davy-United are responsible for over a quarter of this new Indian Steelworks, including all the rolling mills, the wheel and axle plant, the central engineering maintenance shops and the iron and steel foundry. The engineering and supply of all the 75,000 tons of machinery involved is being co-ordinated by the Davy-United Construction Division.



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From 1930 to 1951, apart from his war service, Colonel Mather held various appointments in the locomotive running department and the chief mechanical engineer's department of the LNER. In 1951 he joined The Head Wrightson Machine Company Ltd as assistant general manager, and this year he was appointed managing director.

Colonel Mather served with the Territorial Army from 1926 to 1954. In the Second World War he commanded the 113th L.A.A. Regiment of the Royal Artillery throughout the campaign in Western Europe, and was awarded the O.B.E. (Military) and was mentioned in despatches. He was also awarded the Chevalier Order of Leopold with Palm, and the Croix de Guerre (1940) with Palm. He is now an active member of the North Riding Territorial Army and Auxiliary Forces Association.

Colonel Mather serves on a number of committees and metallurgical societies. He is a member of the committee of the North Eastern Branch of the Institution of Mechanical Engineers, and a member of the Cleveland Institution of Engineers, the Institution of Electrical Engineers, the Institution of Production Engineers, and the British Institute of Management. He is also a member of the American Association of Iron and Steel Engineers.

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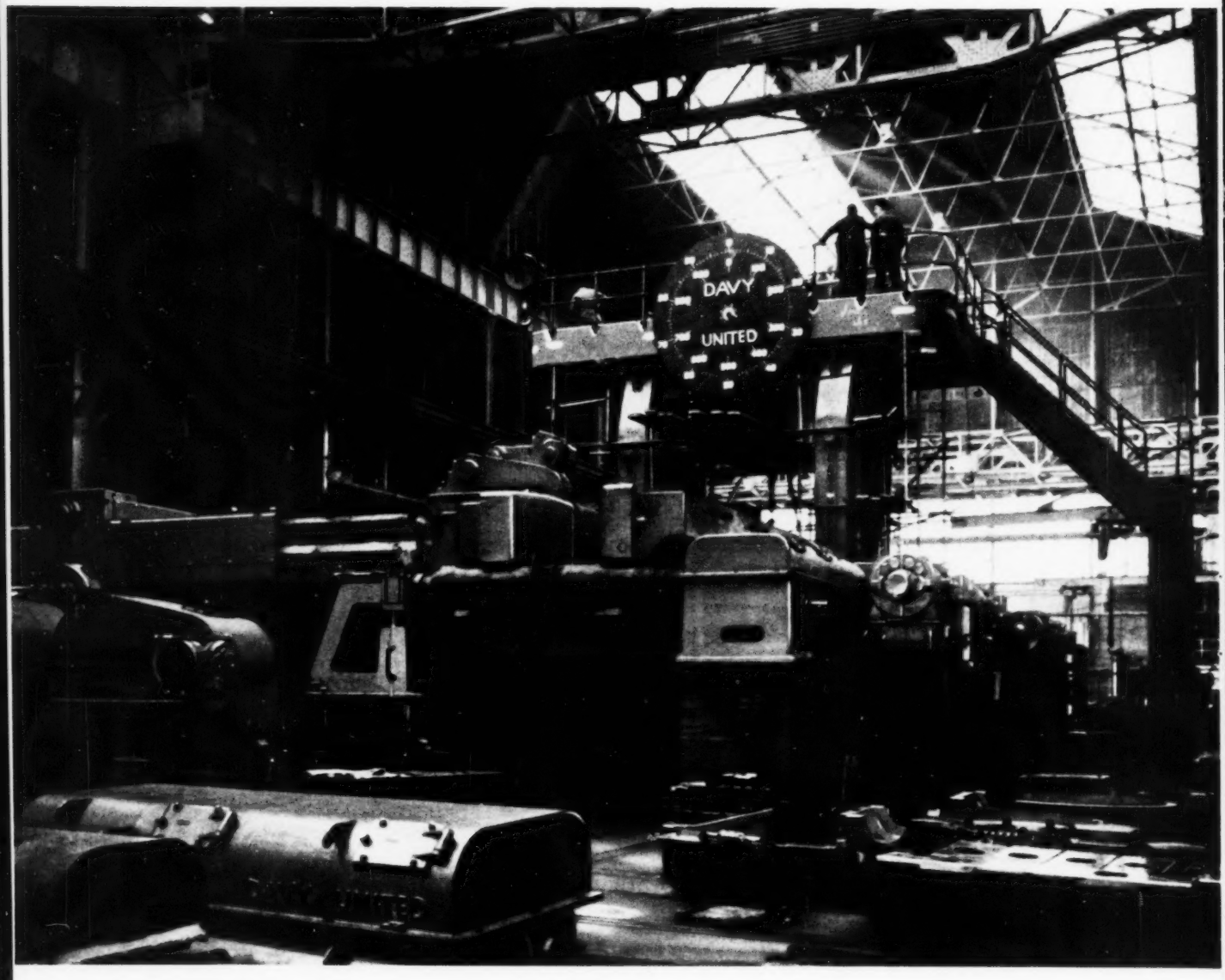

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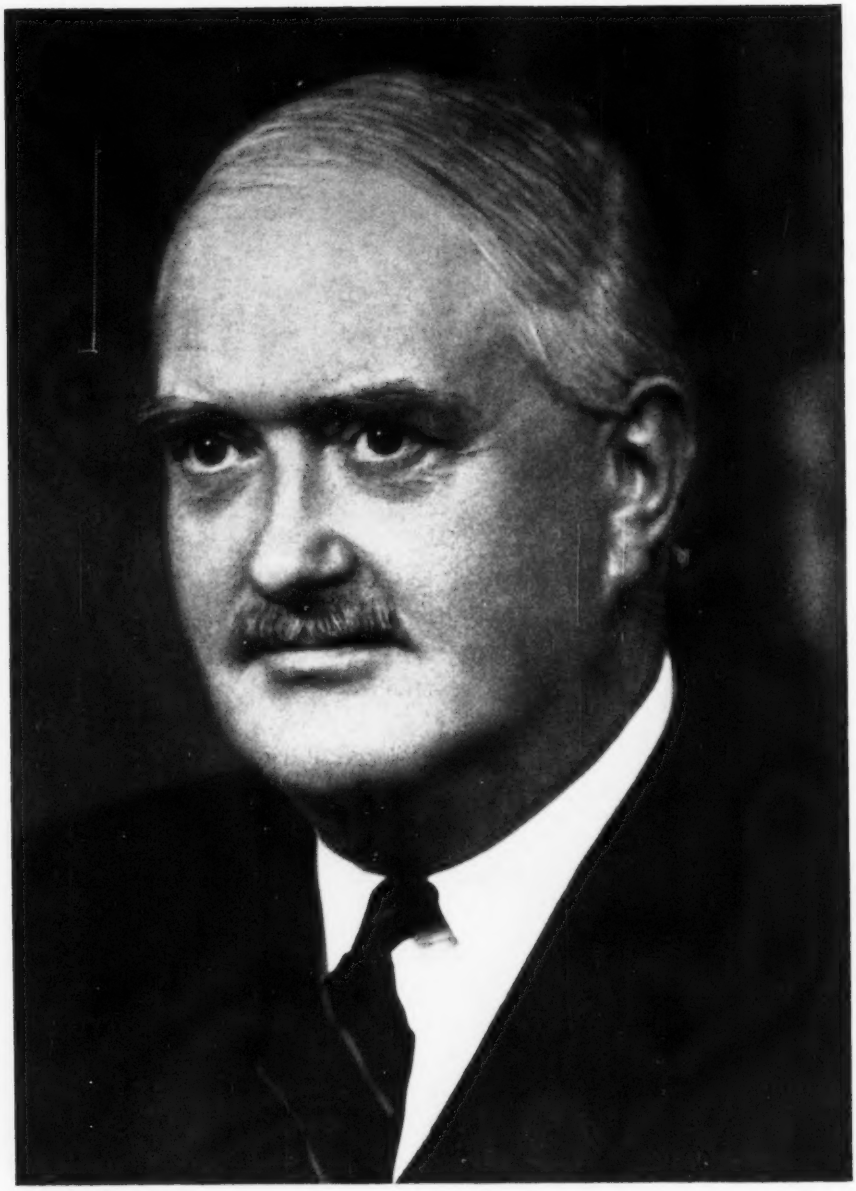


W. H. Mather, O.B.E., T.D., M.I.Mech.E., M.I.E.E.
Honorary Member of Council, 1959–60

Creep, mechanical, and metallurgical properties of an experimental molybdenum–vanadium steel rotor forging

A. I. Smith, B.Sc., A.M.I.Mech.E., A.I.M., E. A. Jenkinson, M.Sc., A.I.M., D. J. Armstrong, A.M.I.Mech.E., and L. M. T. Hopkin, Ph.D., A.R.S.M., A.I.M.

SYNOPSIS

Creep tests have been carried out to obtain design data for a 30-in. dia. turbine rotor in an air-cooled and tempered condition. These tests show that, for designs based on a maximum strain of between 0·1 and 0·3% in a service life of 100 000 h at 3 tons/in², the rotor could operate at a temperature of about 550°C. A few comparative tests on the same rotor in an oil-quenched and tempered condition show a slight superiority in creep resistance at comparatively high stresses, but this is absent at 3 tons/in². Although the short-time tensile properties are somewhat improved by oil-quenching and tempering, as compared with the air-cooled and tempered material, the notched-bar impact transition temperature is also raised. In both conditions of heat-treatment impact resistance is lower at the rim of the rotor than at the bore.

The creep resistance of the rotor even in the quenched and tempered condition is slightly less than that of two other Mo–V *rotors for which data are available. The limited data available suggest that, in general, rotors of* Mo–V *and* 3%Cr–Mo–V–W *steels have similar creep resistances.*

The microstructure of the present rotor consisted of ferrite and pearlite in the air-cooled and tempered condition and ferrite and bainite in the oil-quenched and tempered condition. Extracted residues showed that during creep testing the carbides changed from a predominance of Fe_3C *to that of* V_4C_3 *together with some* Mo_2C. *Chemical analyses of residues showed a significant increase in molybdenum content after creep tests lasting more than 10 000 h at 538° and 600°C so that the test-pieces from which these residues were taken contained less than* 0·25%Mo *in solid solution.* 1853

INTRODUCTION

THE ADOPTION OF STEAM CONDITIONS of 1050°F and 1500 lb/in² and above for the largest power units in the UK has resulted in the use of steels with superior high-temperature properties to those of the carbon and carbon-molybdenum types used for many years. In the case of high-pressure turbine rotors a steel containing about ½%Mo and ¼%V is one of those now used. In consequence, the J/E Subcommittee of the British Electrical and Allied Industries Research Association, which is concerned with the development of steels for steam power plant in the UK, sponsored an investigation, at the National Physical Laboratory, of the creep, mechanical, and metallurgical properties of a 30-in. dia. Mo–V steel rotor.*

The present report includes a brief description of the manufacture of the rotor and its heat-treatment. Creep tests to obtain design data for an estimated life of a power station of 100000 h were made on samples from various parts of the rotor in the air-cooled and tempered condition, this being the condition of heat-treatment in which many rotors are put into service. A few comparative creep tests were made on part of the rotor which had been oil quenched and tempered to find out whether this heat-treatment conferred better creep resistance than air cooling and tempering. The creep properties of the rotor have been compared with limited data available on Mo–V and 3%Cr–Mo–V–W steels used by the Admiralty for smaller diameter rotors for naval purposes.

PARTICULARS OF THE MATERIAL

The rotor forging, of a type suitable for high pressure steam turbines for land power stations, was supplied by Thos. Firth and John Brown Ltd to a specification laid down by the Metropolitan-Vickers Electrical Co. Ltd, who acted as inspecting authority during manufacture. The steel was made by the electric arc process. The specification for the composition of the steel and the cast analysis are given in Table I.

The ingot weighed about 36 tons and was of octagonal cross-section, the distance across the flats at the

Manuscript received 5 July 1960.

Mr Smith and Mr Armstrong are at the Mechanics and Materials Division, National Engineering Laboratory, East Kilbride, Glasgow, and Dr Hopkin and Mr Jenkinson are at the Metallurgy Division, National Physical Laboratory, Teddington.

* Contributions to the cost of the rotor were made by Thos. Firth and John Brown Ltd, The British Thomson-Houston Co. Ltd, The English Electric Co. Ltd, The General Electric Co. Ltd, The Metropolitan-Vickers Electrical Co. Ltd, and C. A. Parsons and Co. Ltd.

TABLE I Cast analysis (wt-%)

	C	Si	Mn	S	P	Mo	V	Ni	Cr
Specification	0·17 to 0·21	0·30 max.	0·4 to 0·6	0·04 max.	0·04 max.	0·6 to 0·75	0·25 to 0·30	0·30 max.	0·30 max.
Cast analysis	0·19	0·28	0·46	0·013	0·010	0·66	0·28	0·36	0·20

top being 56 in. The length of the chill was 7 ft 3 in. The ingot was forged, allowing minimum discards at the top and bottom of 25% and 5% respectively, according to the specification. The barrel of the rotor forging was 32 in. in diameter and 37 in. long while the shaft ends were 15 in. dia. and 37 and 39½ in. long respectively.

The forging was given the following annealing treatment: delivered hot and held at 650°C for 12 h; cooled to 300°C and held for 10 h; reheated to 930–940°C, held for 10 h; cooled to 300°C held for 10 h; heated to 650°C, held for 10 h; and slowly cooled to 100°C.

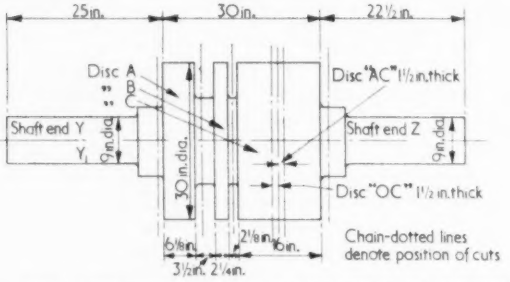
After annealing, the forging was rough machined and gashed to produce three wheels of different thicknesses, as shown in Fig.1, with the exception that material for acceptance tests was left on the body and shaft ends.

The normalizing and tempering treatment given to the rotor was as follows: heated to 950°C and held for 9 h, followed by air cooling to 350°C. The rotor was held at this temperature for 10 h and then reheated to 680°C and held for 8 h. It was then reversed in the furnace from top to bottom and held for a further 6 h at 680°C after which it was furnace cooled to 100°C.

The rotor was sectioned and tested by Thos. Firth and John Brown Ltd. The results of acceptance tests were considered satisfactory by the Metropolitan–Vickers Electrical Co. Ltd. The specification for the acceptance tests is given in Table II.

The shaft ends were then parted off, but the trepanning of the core of the body portion was omitted to hasten the sectioning of the rotor. The rotor was parted to produce three discs *A*, *B*, and *C* as detailed in Fig.1. Sulphur printing and macro-etching were carried out on the transverse faces of all three discs and on diametral sections of discs *A* and *B*. Figure 2 shows a photographic copy of the sulphur prints which were taken on the diametral sections of discs *A* and *B*. The degree of segregation was not considered abnormal.

A disc 1½ in. thick (*AC* in Fig.1) was parted from the mid-section of disc *C* in its air-cooled and tempered condition. The two remaining parts of the body of disc

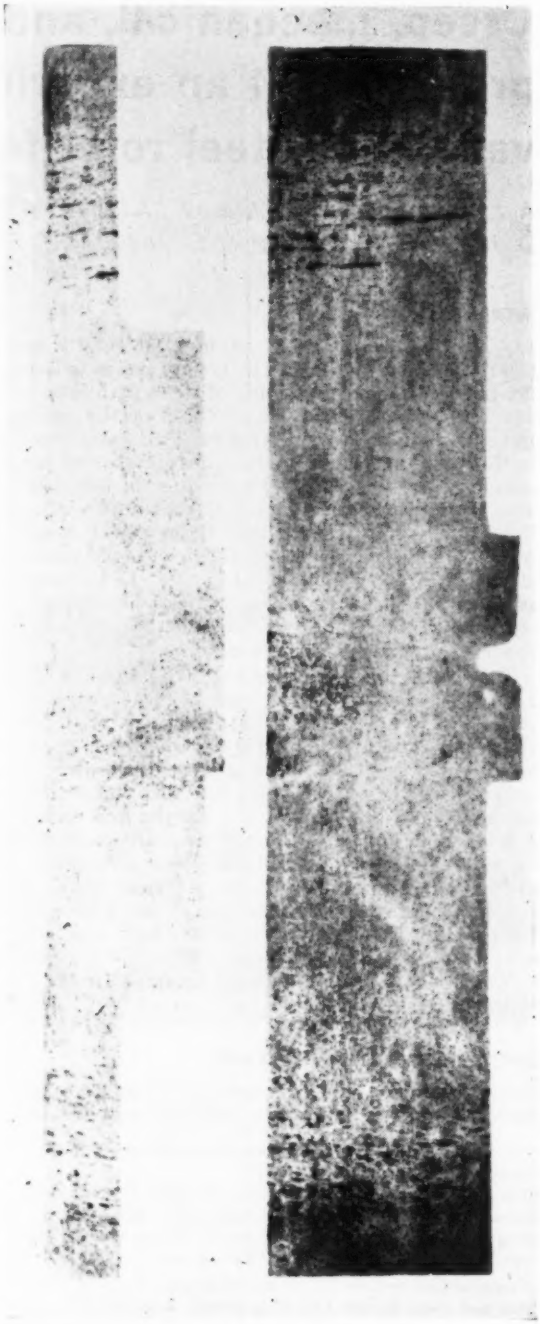


Disc *AC* was parted from disc *C* in the air-cooled and tempered condition, and disc *OC* in the oil-quenched and tempered condition

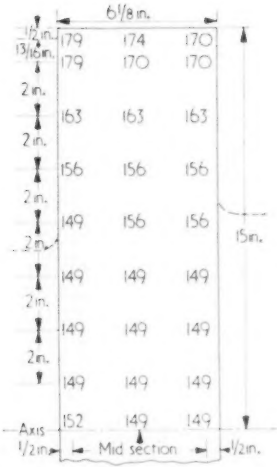
1 *Dimensions and positions of test discs, Mo–V steel rotor forging UBF*

TABLE II Specification for acceptance tests

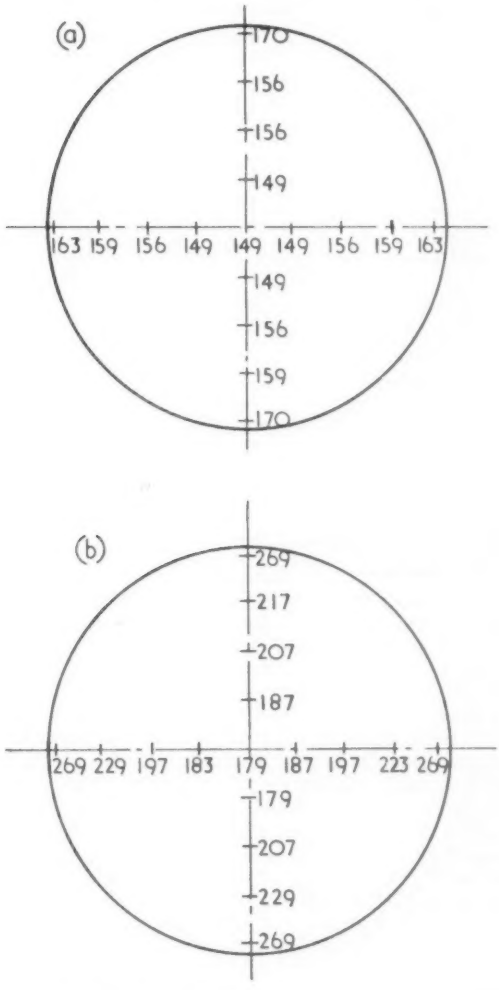
Position	0·2% proof stress, tons/in²	0·5% proof stress, tons/in²	Tensile strength, tons/in²	Elongation, %
Longitudinal	25	Yield point	38	22
Transverse	25	Yield point	38	18
Radial	25	Yield point	38	18



2 *Photographic copy of sulphur prints from surfaces of diametral sections of disc A* (left) *and disc B* (right)



3 *Hardness survey of diametral section of disc A. Brinell hardness numbers (3000 kg load, 10 mm dia. ball)*



(*a*) Air-cooled and tempered
(*b*) Oil-quenched and tempered

4 *Hardness survey on transverse section of disc C 30 in. dia. Brinell hardness numbers (3000 kg load, 10 mm dia. ball)*

C were joined by a peripheral weld, to maintain the original 'mass effect' and the reassembled disc was heat-treated as follows: oil quenched from 970°C, tempered at 680°C for 8 h. After this treatment, a second disc $1\frac{1}{2}$ in. thick (*OC* in Fig.1) was parted from the mid-section of the oil-quenched and tempered disc *C* immediately adjacent to the air-cooled and tempered disc.

Hardness surveys were made on the diametral section of disc *A* and on the transverse faces of the air-cooled and tempered and the oil-quenched and tempered discs from disc *C*. The results are shown in Figs.3 and 4. The discs were then sectioned to obtain test blanks which were forwarded to NPL.

CHEMICAL ANALYSIS

Chemical analyses of samples taken from various parts of the rotor were made at NPL and the results are given in Table III. The analyses of a 4-in. dia. bar forged from the rotor shaft ends[2] has been included for comparison. There is no evidence of marked segregation except that the carbon and molybdenum contents of the rim of each disc are slightly higher than at

TABLE III Chemical analyses of the experimental Mo–V rotor—UBF (no.152) (wt-%)

Element	Disc *A* Rim	Mid-radius	Disc *B* Rim	Bore	Disc *C* Rim	Bore	4 in. dia. bar forged from rotor shaft end
Carbon	0·19	0·14	0·18	0·15	0·20	0·15	0·16
Manganese	0·44	0·44	0·45	0·47			0·46
Silicon	0·24	0·21	0·24	0·23			0·22
Sulphur	0·009	0·008	0·008	0·012			0·015
Phosphorus	0·009	0·008	0·010	0·008			0·009
Nickel	0·43	0·37	0·43	0·38			0·39
Chromium	0·22	0·21	0·23	0·22			0·22
Molybdenum	0·61	0·52	0·61	0·52	0·60	0·56	0·52
Vanadium	0·27	0·26	0·26	0·27			0·27
Tin	0·034	0·036	0·035	0·039			0·035
Aluminium (soluble)	0·001	0·002	0·001	<0·001			0·001
Aluminium (insoluble)	0·003	0·002	0·007	0·004			0·005
Copper	0·15	0·14	0·14	0·14			0·12
Gas analysis (by carbon reduction)							
O_2	0·0036				0·003	0·004	
H_2	Nil				0·00003	0·00002	
N_2	0·010				0·01	0·01	

TABLE IV Results of tensile tests on the air-cooled and tempered material

Tests made in tangential direction. Rate of loading ½ ton/in² min

Position of test-piece	Disc *C* Rim	Disc *A* Rim	Disc *C* Bore	Disc *A* Mid-radius	Disc *A* Mid-radius	Disc *A* Mid-radius	Disc *A* Mid-radius	Disc *A* Mid-radius
Test temperature °C	20	20	20	20	400	500	550	600
Young's modulus, lb/in² × 10⁻⁶	29·0	29·6	29·0	28·5	25·8	23·7	23·4	21·5
Proof stresses, tons/in²								
0·005%	24·0	27·5	20·5	21·7	10·3	10·1	8·5	6·5
0·010%	24·0	27·5	20·5	21·7	10·9	10·6	9·3	7·2
0·050%	24·0	27·5	20·5	21·7	12·7	12·5	11·3	9·0
0·10%	24·6	27·5	20·5	21·7	13·7	13·4	12·2	9·8
0·20%	24·0	27·5	20·5	21·7	15·0	14·5	13·1	10·6
0·50%	24·0	27·5	20·5	21·7	17·2	16·1	14·4	11·7
1·00%	24·0	27·5	20·5	21·7	19·6	17·7	15·6	12·6
Tensile strength, tons/in²	37·1	38·5	32·3	34·1	29·4	24·5	21·4	18·2
Elongation on $4\sqrt{A}$, %	32	30	32*	39	35	39	43	46
Reduction of area, %	63	65	70	72	66	77	82	86

* Broke at gauge mark

the bore position. These analyses substantially agree with the cast analysis made during manufacture except that the molybdenum contents are slightly lower.

MECHANICAL PROPERTIES

Tensile properties

The results of tensile tests made at various temperatures on test-pieces with gauge lengths 2 in. long and 0·357 in. dia. cut from various positions of the rotor are shown for the air-cooled and tempered and the oil-quenched and tempered conditions in Tables IV and V respectively.

Table IV shows that in the air-cooled and tempered condition the properties of the bore positions at 20°C are somewhat lower than for the rim. The values obtained for the mid-radius position of disc *A* are only slightly above those for the bore position of disc *C*, suggesting that the superior tensile properties are confined to the vicinity of the rim only, a conclusion consistent with the hardness distribution (Figs.3 and 4). It will be noted that these tensile test results are, in general, somewhat lower than the values specified for acceptance tests (*cf.* Tables II and IV).

In the oil-quenched and tempered condition of disc *C* there is an even greater difference between the rim and the bore positions (Table V) than in the air-cooled and tempered condition. This is consistent with the greater difference in hardness between the rim and bore positions in the oil-quenched condition (Fig.4). It will be noted that the proof stresses for the rim position of the oil-quenched and tempered material are about 9 tons/in² higher than for the corresponding position of the air-cooled and tempered material, while for the bore position the superiority is about 5 tons/in². It seems that the superior tensile properties of the oil-quenched and tempered material are maintained in tests at 550°C. (*cf.* Tables IV and V. Table IV shows that, at 20°C, the tensile properties of the bore of disc *C* in the air-cooled and tempered condition are less than those for the mid-radius position of disc *A*. This relationship is assumed to remain true at 550°C).

TABLE V Results of tensile tests on the oil-hardened and tempered material

Tests made in tangential direction. Rate of loading, ½ ton/in² min

Position of test-piece	Disc *C* Rim	Disc *C* Bore	Disc *C* Bore
Test temperature, °C	20	20	550
Young's modulus, lb/in² × 10⁻⁶	29·5	29·3	22·2
Proof stresses, tons/in²			
0·005%	32·2	24·3	12·7
0·010%	32·8	25·7	13·4
0·050%	32·8	25·7	15·3
0·10%	32·8	25·7	16·3
0·20%	32·8	26·0	17·4
0·50%	32·8	27·5	18·7
1·00%	33·7	28·0	19·8
Tensile strength, tons/in²	43·7	35·0	24·7
Elongation on $4\sqrt{A}$, %	24	6%* on 2 in.	33
Reduction of area, %	66	68	80

* Broke outside gauge length

Notched-bar impact properties

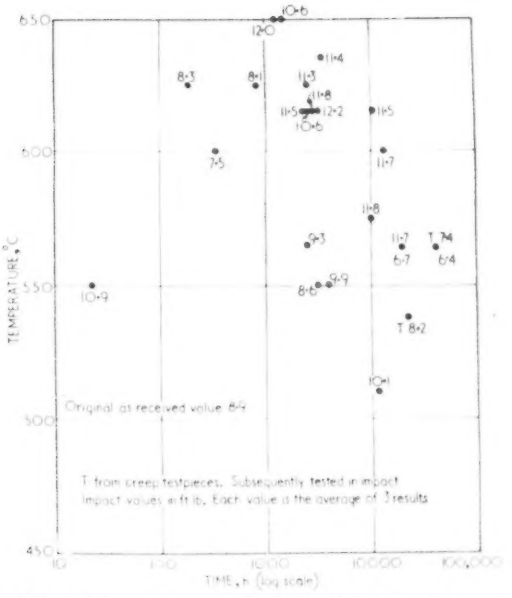
Impact tests on standard 10-mm square three-notch Izod test-pieces cut from various positions of the rotor were made at room temperature and the results are shown in Table VI. In both conditions of heat treatment the material has low Izod values and the fractures are mainly 'crystalline' in appearance. For the bore position the values are somewhat variable but are substantially higher than those for the rim position. The oil-quenching and tempering treatment has lowered the Izod values at the rim position, compared

TABLE VI Results of Izod tests at room temperature

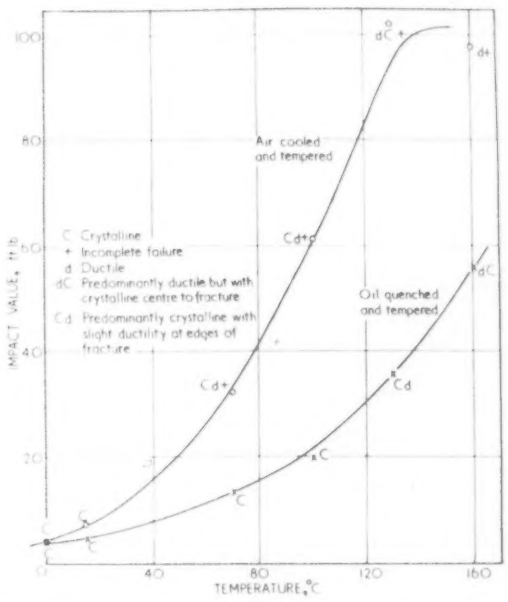
Test-pieces cut with their length in tangential direction. Standard 10-mm sq. three-notch test-piece; 120 ft lb machine

Condition	Position	Izod values, ft lb	Average Izod value, ft lb	Type of fracture
Air cooled and tempered	Rim disc *A*	9, 7, 9	8	Cd
		8, 10, 10	9	Cd
	Mid disc *A*	11, 17, 12	13	Cd
		12, 12, 14	13	Cd
	Bore disc *A*	26, 33	30	Cd
		41, 22, 20	28	Cd
	Rim disc *B*	6, 7, 6	6	Cd
		6, 5, 7	6	Cd
	Rim disc *C*	10, 11, 9	10	Cd
		10, 10, 9	10	Cd
		10, 11, 12	11	Cd
		10, 11½, 12	11	Cd
	Bore disc *C*	21, 41, 10	24	Cd
	Rim disc *C*	4, 5, 7	5	C
Oil quenched and tempered		3½, 5, 5	4½	C
		2½, 3½, 4	3	C
		4, 5, 5	5	C
	Bore disc *C*	28, 14	21	Cd

C = Complete 'crystalline fracture'. Cd = 'Crystalline' with slight ductility at edges of fracture.


TEMPERATURE, °C
TIME, h (log scale)
Original as received value 8·9
T from creep testpieces. Subsequently tested in impact
Impact values in ft lb. Each value is the average of 3 results

5 *Effect of heat-treatment on impact. Resistance of the air-cooled and tempered material taken from the mid-radius position of disc A (Izod test-pieces 5 mm × 3 mm notch tested in 23 ft lb m/c at room temperature)*


IMPACT VALUE, ft lb
TEMPERATURE, °C
Air cooled and tempered
Oil quenched and tempered
C Crystalline
+ Incomplete failure
d Ductile
dC Predominantly ductile but with crystalline centre to fracture
Cd Predominantly crystalline with slight ductility at edges of fracture

6 *Impact properties of air-cooled and tempered and oil-quenched and tempered material from the rim position of disc C*

with the air-cooled and tempered condition, but has not markedly affected the material in the bore position.

Izod tests were also carried out on air-cooled and tempered material after being aged for various times up to 40000 h at temperatures between 510° and 650°C. The 5-mm square Izod test-pieces for these tests were taken from the mid-radius position of disc *A* only. The results of these tests made at room temperature are plotted in Fig.5 where it can be seen that, with ageing, there is no major change of Izod value and there is, therefore, no evidence of significant embrittlement.

The impact properties of the material in both conditions of heat-treatment have been determined over a range of temperatures. Charpy test-pieces of 10-mm square cross-section and 55 mm long with a V-notch cut from the rim position of disc *C*, were used for these tests. The results are shown in Fig.6 and each of the plotted values is the average result of three tests. The diagram indicates that the transition from brittle to ductile fracture for the oil-quenched material occurred at a temperature about 60°C higher than that for the air-cooled material. This result may be related to the tensile strength of the oil-quenched material being higher than for the air-cooled material.[1]

CREEP TEST RESULTS

The creep tests were made, in general, according to B.S.1686 Long-Period, High Sensitivity Tensile Creep Testing. Most test-pieces had gauge lengths 5 in. long and 0·564 in. dia. but for some the gauge lengths were 2 in. long and 0·357 in. dia.

Variation of properties throughout rotor

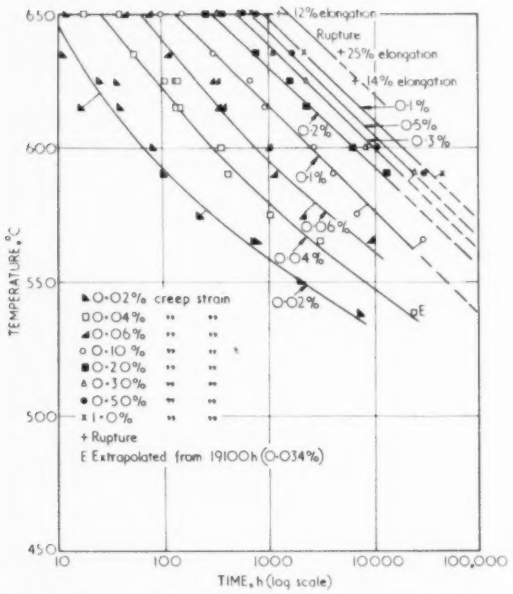
Tests were made at 5 tons/in² and 565°C to determine the variation of creep resistance throughout the rotor. Test-pieces were cut in tangential directions from two positions in each of the discs *A*, *B*, and *C* and, in the case of disc *C*, after various heat-treatments. The results are shown in Table VII where it will be seen that for the air-cooled and tempered material the differences in the time to reach 0·1% strain between the test-pieces taken from various positions of the rotor can be attributed to the normal variability

TABLE VII Tests at 5 tons/in² and 565 C to show variation of creep properties throughout rotor

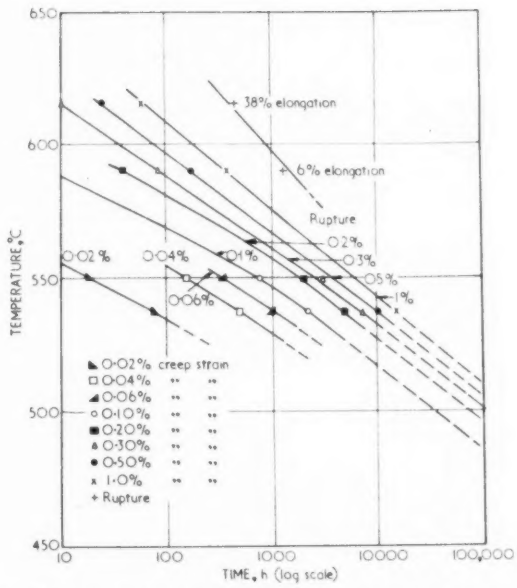
Tests made in tangential direction

Position of test-piece	Heat-treatment	Time, h, to 0·1% strain	0·2% strain	Total duration, h	Final strain, %
Disc *A* Rim	Air cooled and tempered	1770	...	2880	0·13
Disc *A* Mid-radius		1872	5800	20000	0·62
Disc *B* Rim		1705	4850	7654	0·28
Disc *B* Bore		1231	...	2330	0·15
Disc *C* Rim		1532	5225	6310	0·23
Disc *C* Bore		1920	4730	6670	0·26
Disc *C* Rim	Oil quenched and tempered	3850	7386	7386	0·20
Disc *C* Bore		7500E	...	6265	0·09
Disc *C* Rim	Oil quenched and tempered – further 6 h at 680°C	2650	...	4341	0·13
Disc *C* Bore		7000E	...	5566	0·09

E = Extrapolated value



7 *Temperature/log time diagram for specific strains and rupture at 3 tons/in². Air-cooled and tempered material*

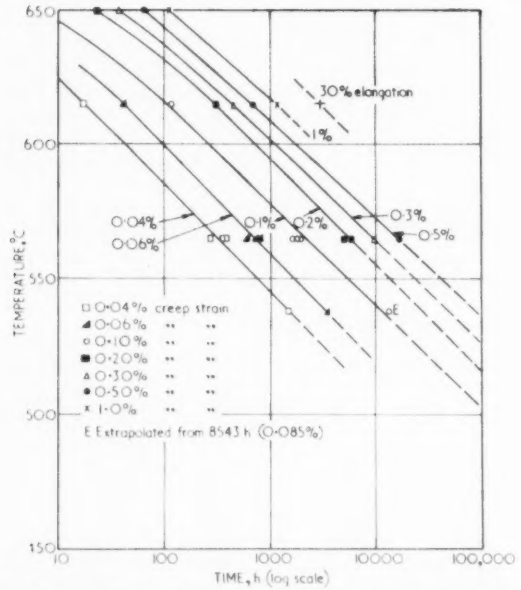


9 *Temperature/log time diagram for specific strains and rupture at 7 tons/in². Air-cooled and tempered material*

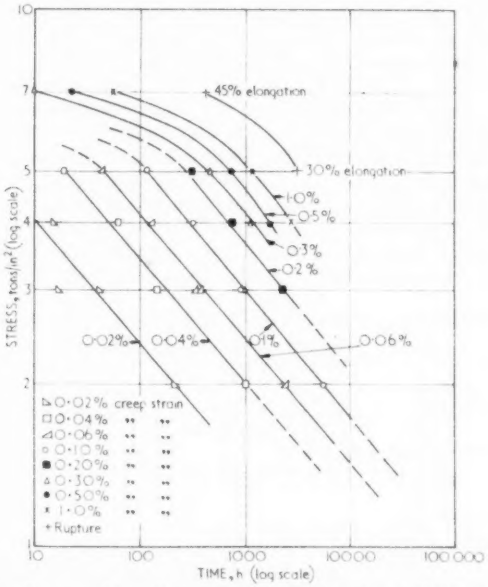
encountered in creep testing. However, in the oil-quenched and tempered condition the material from the bore has a slightly higher creep strength than that from the rim.

Design data for the material in the air-cooled and tempered condition

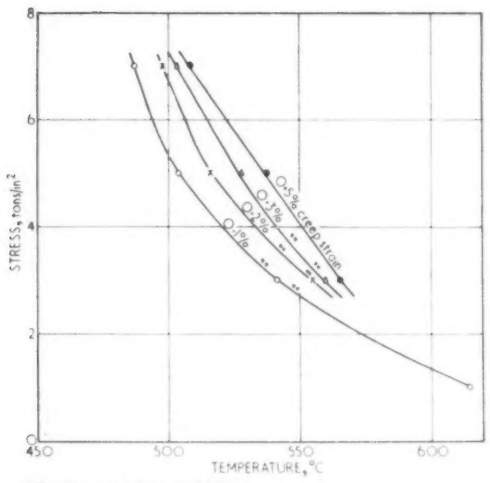
In this section of the investigation creep tests were carried out on material in the air-cooled and tempered condition only, to establish data for use in the design of rotors, i.e. for low strains, in general below 0·5%. In particular it was required to establish the temperatures likely to produce specific creep strains in 100000 h at a stress near to the probable working stress of the material. The main series of tests was begun at 3 tons/in² at temperatures in the range 538 – 650°C and the results are plotted on the temperature/log time diagram, Fig.7. This figure contains results of all the tests at 3 tons/in² on the air-cooled and tempered material including test-pieces cut tangentially from various positions between the bore and rim of the discs *A* and *B*, as well as three test-pieces cut radially



8 *Temperature/log time diagram for specific strains at 5 tons/in². Air-cooled and tempered material*

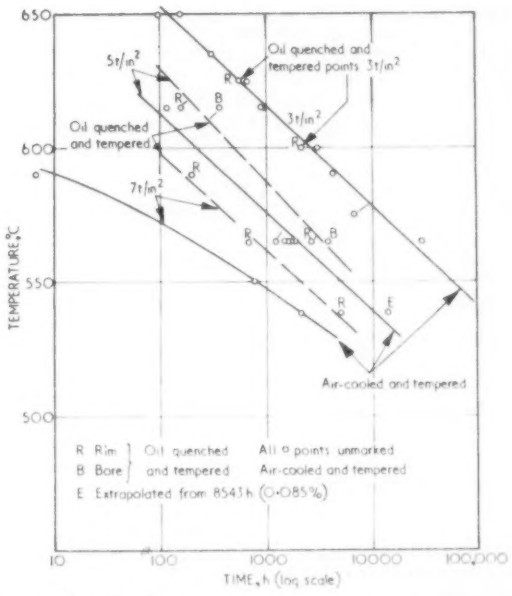


10 *Log stress/time diagram for specified strains at 615°C. Air-cooled and tempered material*



All values are extrapolations

11 *Stress/temperature diagram for specific creep strains in 100000 h. Air-cooled and tempered material*



t/in² = tons/in²

12 *Comparison of creep resistance between air-cooled and tempered, and oil-quenched and tempered materials. Temperature/log time diagram for 0·1% strain at stresses of 3, 5, and 7 tons/in²*

from the bore of disc *A*. The good reproducibility of results is indicative of the high degree of uniformity of the rotor under creep conditions.

Tests were also carried out at 5 and 7 tons/in², and temperature/log time diagrams giving the results are shown in Figs.8 and 9. A few tests were made at various stresses at 615°C. These results are plotted in Fig.10 which shows that the line for 0·1% strain is reasonably well established.

Extrapolation of the lines in Figs.7–10 enables estimates to be made of the creep properties in times of 100000 h. These estimated values are plotted on a stress/temperature diagram in Fig.11.

High strain and rupture characteristics

Examination of the creep curves for the material in both the air-cooled and tempered, and oil-quenched and tempered conditions showed that tertiary creep commenced at low strains, mostly less than 0·2% which were not, apparently, related to either stress or temperature of test. Since the onset of tertiary creep is often an indication of imminent fracture some creep tests were continued to high strains and to fracture. The closeness of iso-strain lines for high strains on log stress/log time and temperature/log time diagrams, gives an indication of the rate of increase of tertiary creep, and the margin of safety which exists for designs based on low creep strains. Moreover, the high strain data facilitates the extrapolation of the lower strain data (which are of more direct use for design purposes) since the iso-strain lines are about parallel.

The high strain data obtained have been included in Figs.7–10. In Fig.7 it will be noted that, for 100000 h at a design stress of 3 tons/in², the temperature to produce 0·3% strain is estimated at about 560°C and for 1% strain at about 570°C, so that a design based on a maximum extension of 0·3% would necessitate a close temperature control.

The results of the tests which continued to rupture are shown in Table VIII. The low elongations obtained with the oil-quenched and tempered material seem less likely to occur with the air-cooled and tempered material. It will be noted, however, that for material in both conditions of heat-treatment, ruptures have not been obtained after sufficiently long times of test to

TABLE VIII Results of rupture tests

Tests made in tangential direction except where stated otherwise

Condition	Position	Stress, tons/in²	Temperature, °C	Time to fracture, h	Elongation on 2 in., %	Reduction of area, %
Air cooled and tempered	Bore disc *A* radial	3	650	1310	>12	43
	Mid-radius disc *A*	3	635	5100	25	43
	Bore disc *A*	3	625	6794	~13	52
	Bore disc *A*	3	615	12112	15	...
	Bore disc *A*	4	615	6859	28*	47
	Mid-radius disc *A*	5	615	3132	30*	59
	Mid-radius disc *A*	7	615	427	45	57
	Mid-radius disc *A*	7	590	1318	6	57
	Rim disc *C*	12	530	2663	10†	53
Oil quenched and tempered	Rim disc *C*	3	625	7416	12	30
	Rim disc *C*	5	615	7051	4	15
	Rim disc *C*	7	590	10446	7	8
	Bore disc *C*	12	530	12942	5	11
	Rim disc *C*	12	530	22569	2	6

* On 1 in. gauge length † Broke outside gauge length

TABLE IX Comparison of the creep strengths at 530°C of the Mo–V power station rotor and admiralty Mo–V and 3%Cr–Mo–V–W rotor steels

Material	Heat treatment	Position	Stress, tons/in²	Creep strain, % 1000 h	10000 h
MoV power station rotor	Air cooled from 950°C and tempered at 680°C	Rim	12	2·14	...
	Oil quenched from 970°C and tempered at 680°C	Rim	12	0·122	0·45
		Bore	12	0·177	...
MoV admiralty rotor*	Water quenched and tempered	Rim	12	0·062	...
		Bore	12	0·108	...
3%Cr MoVW admiralty rotor*	Water or oil quenched and tempered	Rim	12	0·053	...
		Rim	12	0·074	...
		Bore	12	0·082	...
3%Cr MoVW admiralty trial† rotor forgings, forgemaster A		Rim	12	0·097	...
		Bore	12	0·138	...
Forgemaster B	Oil quenched from 1080°C tempered at 670°C	Rim	12	0·115	...
		Bore	12	0·097	...
3%Cr MoVW admiralty rotors†	Oil quenched from 1070°C and tempered at 680°C	Bore	11	0·085	0·186
		Bore	11	0·093	0·176
		Bore	11	0·072	0·182
		Bore	11	0·065	0·134

* Data supplied by steelmakers † Data supplied by Admiralty

reveal minimum values of ductility such as were observed by Glen[9] in a similar steel.

COMPARISON BETWEEN AIR-COOLED AND TEMPERED AND OIL-QUENCHED AND TEMPERED MATERIALS

Comparison of creep resistance between the material in the two main conditions of heat-treatment has been made at stresses of 3, 5, and 7 tons/in² at several temperatures. The results are shown in Fig.12 where the logarithm of the times to reach 0·1% strain are plotted against temperature. It will be noted that, for the rim position, the superiority in creep resistance of the oil-quenched and tempered material observed at 5 and 7 tons/in² does not exist at 3 tons/in².

Figure 12 also suggests that for the oil-quenched and tempered material the creep resistance of test-pieces from the bore position is slightly higher than for the rim. Table VIII shows, however, that the converse is true at 12 tons/in² and 530°C. In the air-cooled and tempered material it has been shown that there is no significant difference in creep resistance of test-pieces from these two positions.

It will be noted in the comparison of creep resistance between the material in the two conditions of heat-treatment in Fig.12 that the air-cooled material was tempered at 680°C for a total time of 14 h whereas the oil-quenched material received only 8 h tempering at this temperature. To determine the effect of this difference in tempering time two test-pieces in the oil-quenched and tempered condition were given a further tempering at 680°C for 6 h and were tested at 5 tons/in² at 565°C. The results of these tests are compared with those of the original oil-quenched and tempered condition of heat-treatment in Table VII; the additional tempering time has only slightly altered the time to reach 0·1% strain.

COMPARISON WITH OTHER ROTOR STEELS

Creep data on a large $\frac{1}{2}$%Mo–$\frac{1}{4}$%V rotor forging have been reported by Kirkby[3,4] who presented his results as a stress/log time diagram for specific creep strains at 538°C for times up to 30000 h. Comparison of these results with those obtained on the present rotor suggests that in both conditions of heat-treatment the present material is somewhat inferior to that tested by Kirkby. The rotor tested by Kirkby[5] was 28 in. dia. and was oil-quenched from 960°C and tempered at 680–690°C for 13 h. Its composition and mechanical properties were as follows:

C	Si	Mn	S	P	Ni	Cr	Mo	V
0·21	0·26	0·53	0·012	0·009	0·36	0·35	0·75	0·28%

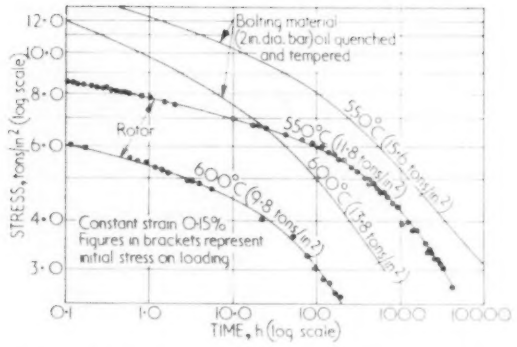
	0·2%PS, tons/in²	TS, tons/in²	El$_5$,%	R of A,%
Body end (tangential direction)	50·2	55·6	20	55
Body end (radial direction)	48·2	53·8	21·5	60

It will be noted that the size and heat-treatment is similar to that for the oil-quenched and tempered portion of the present rotor. The chromium and molybdenum contents of Kirkby's rotor, however, are slightly higher which may account for its higher creep and short-time tensile properties.

A few tests on Mo–V forgings of substantial size have also been reported by Kirkby and Sykes,[6] and Burton, Russell, and Walker,[7] but these were carried out under testing conditions not directly comparable with the present tests.

Cowlin and Veitch[8] have described the post-war development of high-temperature steam turbines for naval use. One of the creep design requirements for HP rotor material specifications was that 0·1% creep strain should not be exceeded in 500 h at 12 tons/in² and 530°C; conditions representing the maximum stressing of the rotor during full power in service. Two of the steels considered for use as rotors were Mo–V, water-quenched and tempered, and 3%Cr–Mo–V–W, and some creep test results on several rotor forgings of these materials, at 12 tons/in² and 530°C have been made available by the steelmakers and the Admiralty. Although these test conditions are probably not representative of operating conditions in large power-station turbines, comparison of the data from the steelmakers and the Admiralty with the results of similar tests on the present rotor is of some interest.

Creep tests were, therefore, made at 12 tons/in² and 530°C on test-pieces cut from disc *C* in both the air-cooled and tempered, and oil-quenched and tempered conditions. The results are reproduced in Table IX together with those for several Admiralty rotor forgings. The test on the present rotor in the air-cooled and tempered condition showed very rapid creep as



13 *Curves for stress-relaxation tests on the air-cooled and tempered material (mid-radius of disc C) and on a* Mo–V *bolting material*

might be expected in view of the relatively low proof and tensile stresses of this material compared with that in the oil-quenched and tempered condition (*cf.* Tables IV and V). Test-pieces from the rim and bore positions of the rotor in the oil-quenched and tempered condition showed substantially higher creep resistance but this was still not quite as good as that reported by the steelmakers on the water-quenched and tempered Mo–V Admiralty rotor. Undoubtedly the smaller diameter of the Admiralty rotor and the water-quenching treatment enables more drastic quenching to be carried out than in the case of the present oil-quenched rotor. The improvement of the proof stresses by drastic quenching would be expected to result in improved creep resistance at high stresses and low temperatures but the results of the present tests suggest that improved creep resistance obtained in this way may not exist at low stresses.

It will be noted from Table IX that for rotors in the quenched and tempered condition the Mo–V and 3%Cr–Mo–V–W steels have similar creep resistances. The composition and mechanical properties of the 3%Cr–Mo–V–W steel rotors for which data were supplied by the Admiralty are as follows:

C	Si	Mn	S	P
0·19–0·21	0·26–0·32	0·41–0·49	0·005–0·007	0·010–0·013
Ni	**Cr**	**Mo**	**V**	**W**
0·35–0·43	2·58–2·68	0·68–0·79	0·75–0·88	0·57–0·74

	0·2%PS, tons/in²	TS, tons/in²	El., %	R of A, %
Radial, body	51·1–52·2	59·6–61·2	18·0–19·0	45·0–50·0
Tangential, body	51·8–53·6	60·0–62·4	13·0–18·0	25·0–40·0

Table IX also shows that there is no consistent difference in creep resistance between test-pieces from the rim and bore positions of the rotors.

COMPARISON WITH OTHER MO–V STEELS

Creep and rupture tests

Fairly extensive creep data on Mo–V steel in two other forms have already been reported by Smith, Jenkinson, Armstrong, and Day.[2] Both steels were nominally $\frac{1}{2}$%Mo, $\frac{1}{4}$%V but differed, in detail, in analysis and heat-treatment from the rotor steel discussed in this paper; one was in the form of a steam-pipe and the other a cast slab representative of castings for turbine casings, steam chests, etc. Comparison of the creep properties of these three steels indicates that the air-cooled and tempered rotor and the casting are about the same and the pipe steel is 10–20°C superior.

Data on the high temperature properties of Mo–V steel have also been published by Glen.[9] Most of Glen's numerous creep tests were on bar material of small diameter but many were continued to rupture. The creep properties reported by Glen appear to be rather better than those of the steel of the present rotor and similar to those for the pipe material.

Relaxation tests

Two relaxation tests were made on the air-cooled and tempered material adopting the method described by Herbert and Armstrong.[10] In this method of stress-relaxation testing the results are independent of elastic 'follow up' of any components stressed in series with the test-piece. These tests were made to ensure that no abnormal behaviour occurred under relaxation conditions which may exist at stress-concentrations, e.g. at blade root fixings. The results are shown in Fig.13 compared with those for oil-quenched and tempered bolting material (2 in. dia. bar) of the same composition.[2] The test-pieces were initially stressed to produce 0·15% strain in each case. Figure 13 shows that the rotor material has lower resistance to relaxation than the bolting material; this probably being associated with the lower proof stresses of the rotor material.

METALLURGICAL EXAMINATION OF THE ROTOR FORGING

Microstructure

Discs *A* and *B* had been air-cooled and tempered; the structures consisted of pearlite and ferrite areas, the latter containing a small proportion of precipitated carbide (Figs.14*a–d*). The carbide in the pearlite areas was somewhat spheroidized, though in a few places lamellar pearlite was present. A large inclusion was encountered in the bore position of disc *B*; the structure at this inclusion is shown in Fig.14*b*.

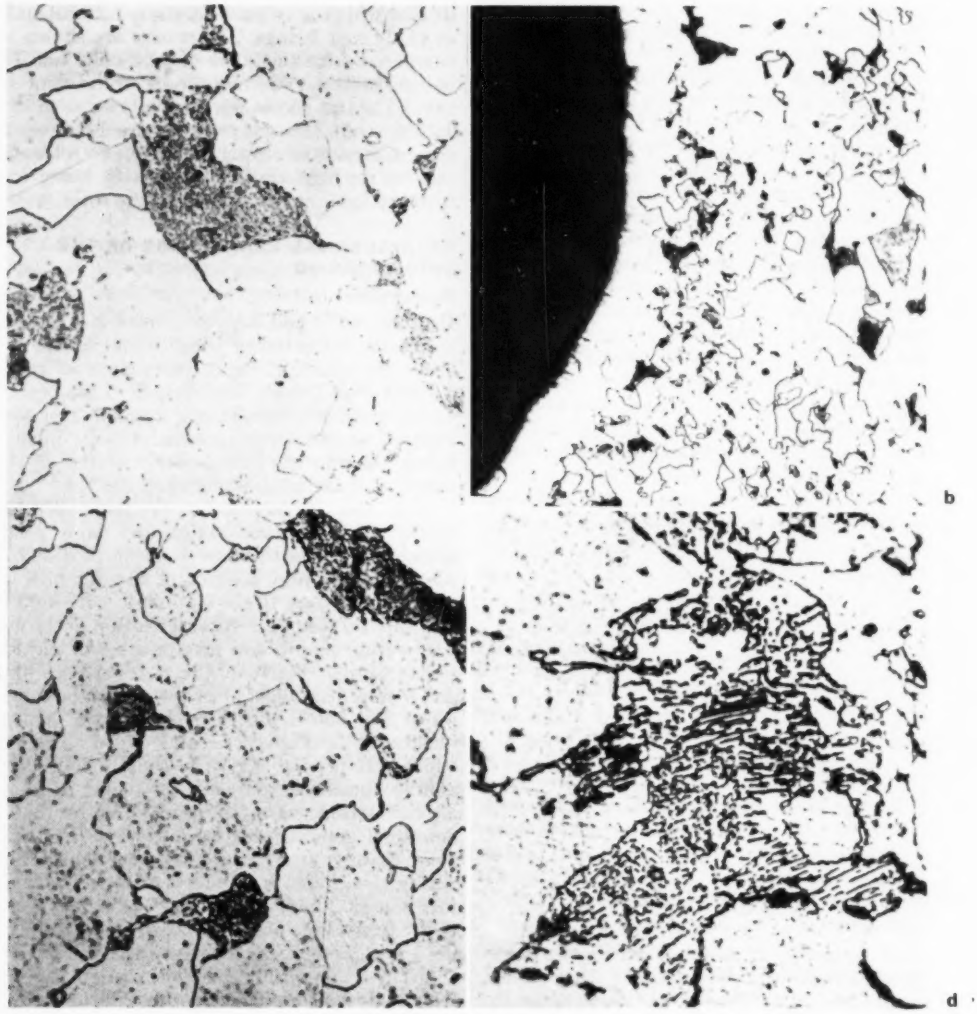
A number of creep test-pieces have been examined chiefly from the mid-radius and bore positions of disc *A*, and from the bore position of disc *B*. A test-piece from the bore position of disc *B* which had been tested in creep for 19110 h at 538°C with 0·03% extension showed complete spheroidization of the carbide in the pearlite grains, and precipitation of carbide in the ferrite grains (Fig.16*b*). Creep test-pieces of discs *A* and *C* were examined after testing in the temperature range 530–650°C and the changes in structure are illustrated in Figs.16*b–d* and 17*a–d*. The test-piece which fractured at 530°C in 2663 h showed considerable elongation of grains in the region of fracture; recrystallization was also proceeding in these highly deformed areas. Holes were present (Fig.16*a*) which were mainly at grain boundaries, but a few were within the grains. In the tests at higher temperatures, for example, those at 590° and 615°C, recrystallization of the ferrite was marked and widened grain boundaries were present in the test-piece tested at 600°C (Fig.17*a–d*). In all test-pieces the original pearlite and ferrite grains altered considerably during creep testing and precipitation of additional carbide was noted particularly in the structure of test-pieces tested at the higher temperatures.

Portions of disc *C* were examined before test in the air-cooled and tempered, oil-quenched and tempered, and also in the oil-quenched and tempered condition with additional tempering. The structures at the rim and bore positions are illustrated for the three heat-treatments in Figs.14*c–d* and 15*a–d*. The air-cooled and tempered samples of disc *C* showed similar structures at the rim and bore positions consisting of lamellar pearlite areas with ferritic areas containing a precipitated carbide. These structures contained more lamellar pearlite than was found in discs *A* and *B*. The oil-quenched and tempered samples of disc *C* showed bainitic and ferritic areas with some precipitated carbide in the ferritic areas. The bainite in the bore position appeared less well resolved than the bainite occurring in the rim. The effect of tempering for a further period of 6 h at 680°C produced greater spheroidization in all positions.

A creep test-piece in the oil-quenched and tempered condition showed, after 7386 h at 565°C, well resolved carbide in the bainitic areas, and a fine precipitated carbide in the ferrite grains (Fig.18*a*). No cracking was observed in this test-piece; the final creep strain being 0·20%. Two other samples of the oil-quenched and tempered material of disc *C* after creep tests at 615°C were examined; one at the bore position and one at the rim. In these examples the bainitic areas had nearly disappeared and spheroidized carbides were found in the ferrite grains (Figs.18*b* and *c*). The test-piece fractured at 615°C contained intercrystalline cracks at the test length and oxide penetration had taken place along grain boundaries. A further test-piece, which ruptured after 7416 h at 625°C had structures similar to those shown in the test-pieces ruptured at 615°C (Fig.18*d*).

X-ray examination of residues

X-ray examinations were made of residues extracted from the specimens by solution of the samples in an alcoholic solution of iodine. Details are given in



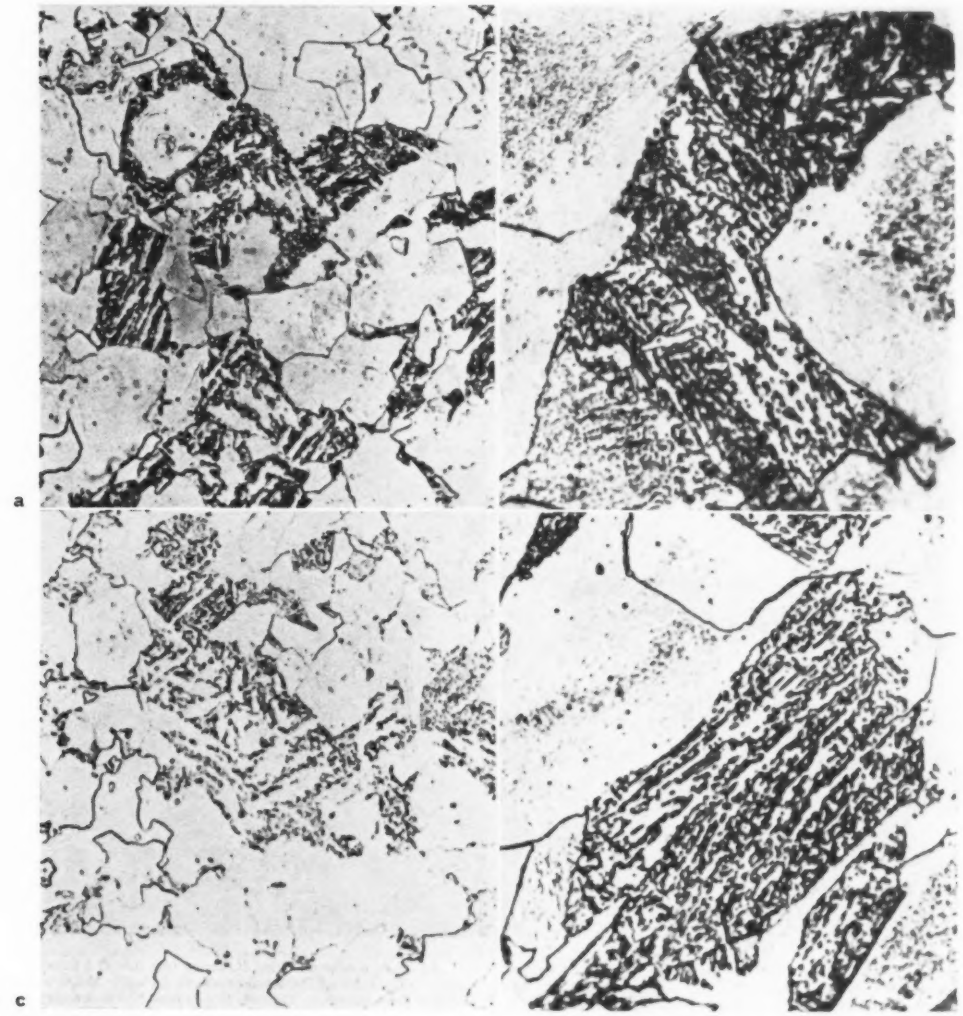
(*a*) Rim of disc *A*. Carbides—$Fe_3C + V_4C_3$ ×500
(*b*) Bore position of disc *B*. Inclusion shown on left ×100
(*c*) Bore position of disc C. Carbides—$Fe_3C + V_4C_3$ ×500
(*d*) Rim position of disc *C* ×1500

14 *Discs, A, B, and C (air-cooled and tempered material). All specimens etched in a solution of 4% nitric acid in alcohol*

TABLE X Results of X-ray examination of residues extracted from rotor specimens of steel UBF (no.152)

Position	Condition, creep test results	Residue content, %	Carbides identified	Hardness (DPN)
Air-cooled and tempered				
Rim of disc *A*	Before test	2·52	$Fe_3C + V_4C_3$ weak	...
Rim of disc *A*	Before test	2·50	$Fe_3C + V_4C_3$	196
Rim of disc *C*	Before test	2·25	$Fe_3C + V_4C_3$	168
Bore of disc *A*	Before test	1·64	$Fe_3C + V_4C_3$	154
Bore of disc *C*	Before test	1·31	$Fe_3C + V_4C_3$	159
Bore of disc *B*	538°C, 3 tons/in², 19106 h (U), 0·03% elongation	1·50	$V_4C_3 + Fe_3C + Mo_2C$ weak	146
Bore of disc *A*	575°C, 3 tons/in², 10242 h (U)	2·76	$V_4C_3 + Fe_3C$ weak $+ Mo_2C$ weak	143
Bore of disc *A* (radial)	600°C, 3 tons/in², 12705 h (U), 0·7% elongation	0·82	$V_4C_3 + Mo_2C$ weak	124
Mid-radius of disc *A*	615°C, 5 tons/in², 2370 h (U)	1·34	$V_4C_3 + Fe_3C$ weak $+ Mo_2C$ weak	142
Mid-radius of disc *A*	615°C, 4 tons/in², 3211 h (U)	2·20	$Fe_3C + V_4C_3 + Mo_2C$ weak	139
Mid-radius of disc *A*	650°C, 5 tons/in², 242 h (U), 3·74% elongation	1·27	$V_4C_3 + Fe_3C + Mo_2C$ a little	144
Oil-quenched and tempered				
Rim of disc *C*	Before test	2·90	Fe_3C	256
Bore of disc *C*	Before test	1·88	Fe_3C	193
Rim of disc *C*	As above + 6 h 680°C	2·94	$Fe_3C + V_4C_3$ trace	251
Bore of disc *C*	As above + 6 h 680°C	1·97	$Fe_3C + V_4C_3$	187
2 in. dia. bar forged from rotor shaft	Before test	1·83	$Fe_3C + V_4C_3$	214
2 in. dia. bar forged from rotor shaft, Izod blank	22578 h, 525°C unstressed	2·20	$Fe_3C + V_4C_3 + Mo_2C$ weak	227
2 in. dia. bar forged from rotor shaft, stressed testpiece	22578 h, 525°C, stressed	1·64	$Fe_3C + V_4C_3 + Mo_2C$ probably	238

U = Unbroken



(*a*) Rim of disc *C* (oil quenched and tempered). Carbide—Fe_3C ×500

(*b*) Bore of disc *C* (oil quenched and tempered). Carbide—Fe_3C ×500

(*c*) Rim of disc *C* (oil quenched and tempered + 6 h 680°C). Carbides—$Fe_3C + V_4C_3$ trace ×500

(*d*) Bore position of disc *C* (oil quenched and tempered + 6 h 680°C). Carbides—$Fe_3C + V_4C_3$ ×1500

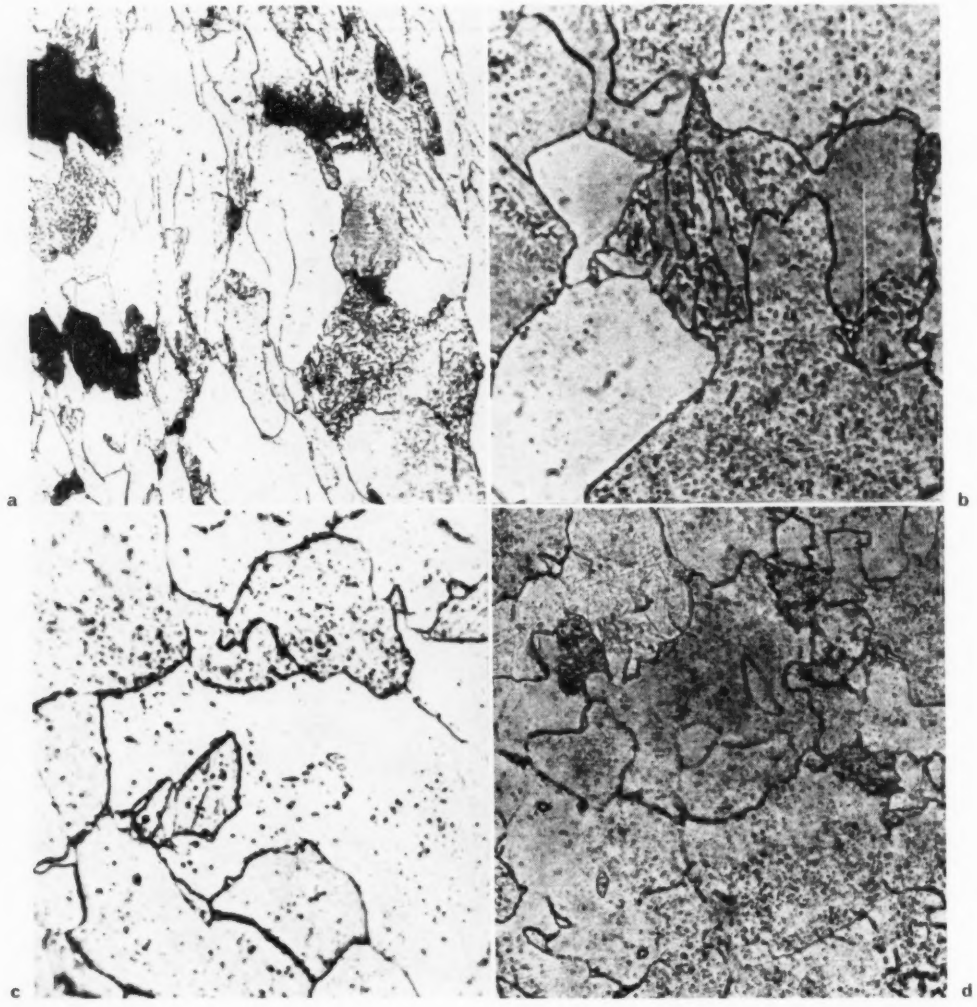
15 *Disc C (oil-quenched and tempered material). All specimens etched in a solution of 4% nitric acid in alcohol*

TABLE XI Chemical analyses for Mo in extracted residues from rotor specimens of steel UBF (no.152)
Mo content of steel, 0·52–0·60%

Position in disc, etc.	Condition, creep test results, etc.	Residue, %	%Mo in steel found in residue	% Mo content of sample,	%Mo in matrix (by difference)
Bore of disc *C*	Before test (air cooled and tempered)	1·31	0·19	0·56	0·37
Bore of disc *B*	Air cooled and tempered; 19 106 h, 538°C, 3 tons/in², 0·03% elongation	1·50	0·33	0·52	0·19
Bore of disc *A* (radial)	Air cooled and tempered; 12 705 h, 600°C, 3 tons/in², 0·7% elongation	0·82	0·30	0·52	0·22
Bore of disc *C*	Before test (oil quenched and tempered)	1·88	0·20	0·56	0·36
Rim of disc *C*	As above + 6 h 680°C	2·94	0·33	0·60	0·27
Bore of disc *C*	As above + 6 h 680°C	1·97	0·30	0·56	0·26

Table X. A few chemical analyses for molybdenum were made in certain samples and the results are reported in Table XI. The carbides identified are given in Table X.

In the air-cooled and tempered condition, the disc *A* at the rim and bore positions contained the carbides Fe_3C and V_4C_3. During creep tests at 530–650°C (discs *A*, *B*, and *C*) the carbide Fe_3C decreased in quantity and V_4C_3 became the prominent carbide; with long times of testing small proportions of Mo_2C were detected. These results can be compared with those obtained for the 4 in. dia. forged bolt steel pre-



(*a*) Close to fracture. Rim disc *C*. Creep test 530°C 12 tons/in² 2663 h El. 10% on 2 in. ×500
(*b*) Bore positions. Disc *B*. Creep test 538°C 3 tons/in² 19 110 h (U). 0·03% El. Carbides—V_4C_3 + Fe_3C + Mo_2C weak ×1500
(*c*) Disc *A*. Creep test 565°C 3 tons/in² 40 206 h (U). 0·12% El. ×1500
(*d*) Bore position. Disc *A*. Creep test 575°C 3 tons/in² 10 242 h (U). 0·11% El. Carbides—V_4C_3 + Fe_3C (weak) + Mo_2C (weak) ×500

16 *Discs A, B, and C (air-cooled and tempered material). All specimens etched in a solution of 4% nitric acid in alcohol*

pared from the rotor shaft also given in the table. For the bolt steel, Fe_3C was the dominant carbide although the quantity of V_4C_3 increased with time of testing.

Samples of disc *C* in the oil-quenched and tempered condition, and also after additional tempering for 6 h at 680°C, were examined, and showed Fe_3C as the main carbide; in the samples which had received the extra tempering small proportions of V_4C_3 were also present.

The analyses for molybdenum in the residues show that there was a considerable increase in the molybdenum content during creep testing; for example, exposure to 538°C in 19106 h. increased the molybdenum content of the residue from 0·19 to 0·33%, and by difference the content of the matrix decreased from 0·37 to 0·19%. Similarly during creep testing at 600°C for 12705 h the molybdenum content of the residual rose from 0·19 to 0·30%, the content of the matrix fell from 0·37 to 0·22%.

In the oil-quenched and tempered samples the molybdenum contents of the residue and matrix were similar to those for the air-cooled and tempered samples, but the residues from the oil-quenched and tempered samples which received additional tempering at 680°C contained an increased proportion of molybdenum.

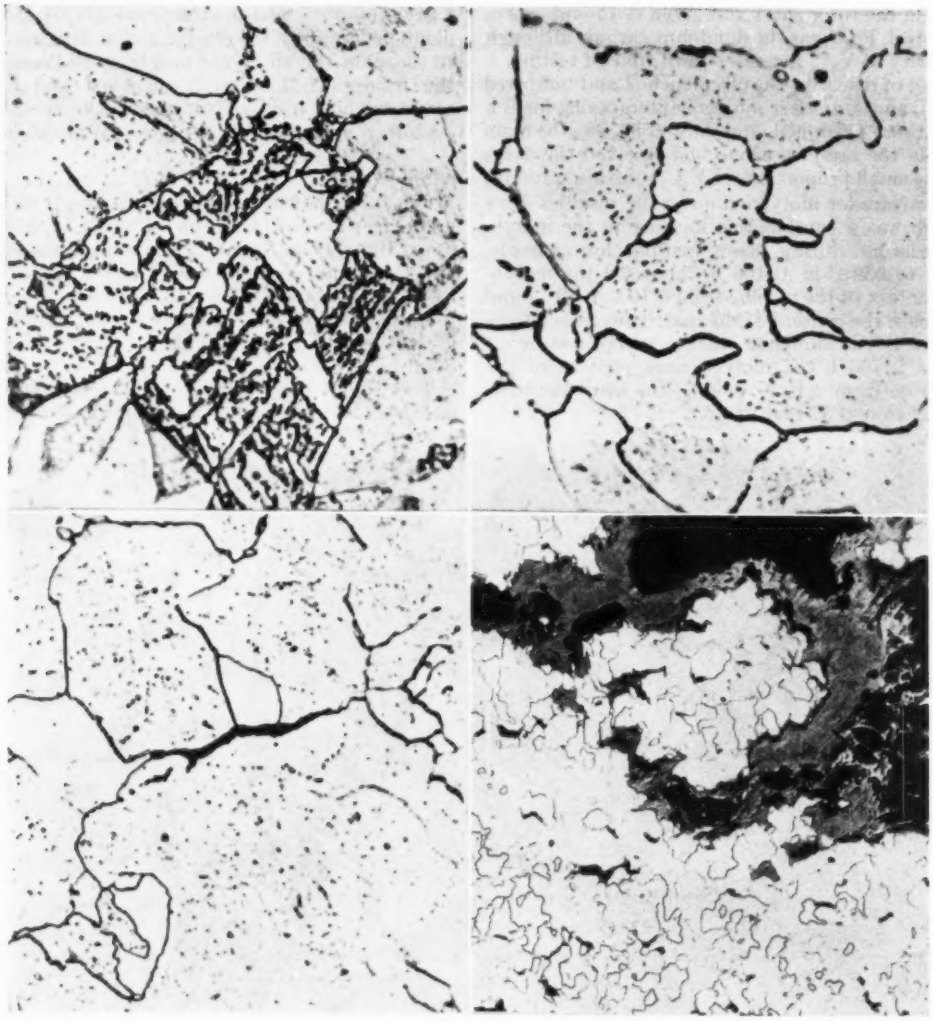
CONCLUSIONS

It has been shown that for designs based on a maximum creep strain of between 0·1 and 0·3% in a service life of 100000 h at 3 tons/in² the rotor in the air-cooled and tempered condition would operate at a temperature of about 550°C. Slight increases in operating temperature above this value would rapidly cause greater creep strains, e.g. at 570°C, 1% creep would occur in 100000 h.

The slight superiority of the rotor in the oil-quenched and tempered condition, compared with the

(*a*) Mid-radius. Disc *A* section close to fracture. Creep test 590°C, 7 tons/in² 1318 h (B), 5·5%El. ×500
(*b*) Bore position. Disc *A* creep test 600°C, 3 tons/in² 12700 h (U), 0·7%El. Carbides—V_4C_3 + Mo_2C (weak) ×1500
(*c*) Mid-radius. Disc *A* creep test 615°C, 4 tons/in² 3211 h (U), 1·3%El. Carbides—Fe_3C + V_4C_3 + Mo_2C (weak) ×500
(*d*) Mid-radius. Disc *A*. Section close to fracture. Creep test 615°C, 7 tons/in² 427 h (B), 45%El. ×1500

17 *Disc A (air-cooled and tempered material). All specimens etched in a solution of 4% nitric acid in alcohol*



(*a*) Rim. Disc *C*. Creep test 565°C, 5 tons/in² 7386 h (U), 0·2%El. ×1500
(*b*) Bore position. Disc *C* creep test 615°C, 5 tons/in² 3475 h (U), 1·0%El. ×1500
(*c*) Rim. Disc *C*. Close to fracture. Creep test 615°C, 5 tons/in² 7051 h (B), 3·8%El. ×1500
(*d*) Bore position. Disc *C* at fracture. Creep test 625°C, 3 tons/in² 7416 h (B), 11·5%El. ×150

18 *Disc C (oil-quenched and tempered material). All specimens etched in a solution of 4% nitric acid in alcohol*

air-cooled and tempered condition, may not be maintained at 550°C throughout the service life of such a rotor.

The limited data available for the 3%Cr–Mo–V–W rotor steel suggest that this material has similar creep resistance to the Mo–V rotor steel when both are in the quenched and tempered condition.

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The electrolytic production of sulphur-free iron

W. T. L. Neal, M.A., J. F. Duke, B.A., and H. R. Sullivan

SYNOPSIS

A patented process for the production of iron with a sulphur content below 5 ppm is described. The method involves electrolysis across an ion exchange membrane separating a two-compartment cell, and is capable of producing 500–600 g of iron in 7 h. 1841

THE PRONOUNCED EFFECT of small concentrations of non-metallic elements such as carbon, oxygen, and sulphur on the physical and mechanical properties of pure iron is well known, and has led to many attempts to reduce the concentrations of these elements to the lowest possible level.

Workers in the Metallurgy Division of the National Physical Laboratory have achieved considerable success in this direction using pyrometallurgical techniques. A typical sequence of operations on the 25-lb scale has been described by Hopkins[1] which results in a material of which the non-metallic impurity concentrations fall in the ranges shown below.

0·002–0·006%S; 0·001–0·002%P; 0·001–0·004%C;
0·0005–0·001%N; 0·0005–0·0002%O; <0·000005%H

As a result of these operations it became apparent that it was not practicable to lower the sulphur content further by variations of the process and hence an alternative method was sought whereby the sulphur concentration could be reduced to less than 0·0005%.

Moore[2] has discussed various methods available for the purification of iron. The method selected must, of course, depend upon the elements to be removed, and its applicability to the scale of production required. At the US Bureau of Standards, Moore has used a method based on the ether extraction of ferric chloride from an aqueous solution. While such a procedure is potentially capable of producing sulphur-free iron, the difficulties, hazards, and tedium involved in the production of, say, 100 kg of iron a week by this method do not make it very attractive. Many of the other processes discussed by Moore in his paper suffer similarly from the drawback of inconvenience and cost when expanded to the scale of production projected in this laboratory. The only process thought likely to be practicable was an electrolytic one, and this paper outlines the development of an electrolytic method[3] whereby iron with a sulphur content of less than 0·0005% has been obtained. Details of the main requirements of the equipment necessary to produce 10 kg of iron a week are given.

PRELIMINARY WORK

After a little exploratory work, a cell was set up in a rectangular glass tank to the following specifications:

Cell dimensions	15 cm × 20 cm × 30 cm deep
Volume of electrolyte	5 l.
Composition of electrolyte	100 g Fe/l. as $FeCl_2$ in HCl at pH = 2
No. of anodes	6
Area of each anode (submerged)	175 cm^2
No. of cathodes	5
Area of each cathode	190 cm^2
Current density	2·36 Adm^{-2}

The anodes and cathodes were arranged alternately, and the cell stirred by slowly oscillating the anodes. In a typical run a total of 8·5 kg of anode iron was dissolved and 6·3 kg of iron firmly deposited on the cathodes. The cathodes were removed at intervals during the deposition and the sulphur content of the deposited iron determined with the results shown below.

Deposit no.	Mass deposited, g	S, %
1	75	0·0020
2	550	0·0030
3	800	0·0035
4	3300	0·0035
5	1600	0·0035

The anode iron contained 0·013% of sulphur, and it was thought reasonable to suppose that the value of 0·0035% for the sulphur content of the electrodeposited iron is the equilibrium value with the particular anode material used, some of the sulphur in the anode iron being transferred to the cathode.

Consequently, it was necessary to devise a method to remove sulphur from the electrolyte, and the following ways of achieving this were contemplated:

(i) chemical precipitation of the sulphur, coupled with filtration or centrifugation
(ii) removal of sulphur on ion-exchange columns
(iii) use of a two-compartment cell fitted with a diaphragm which would hinder the diffusion of sulphur-containing particles or ions towards the cathode.

Method (i) was discarded on the ground that sulphur was probably present in a variety of ionic forms, and that the precipitation and separation of such trace

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The authors were at the National Physical Laboratory, Teddington, Middlesex, when this paper was written. Mr Neal is now with the Ministry of Housing and Local Government, London.

amounts on a continuous basis would be very difficult, if not impossible. Likewise, method (ii) had to be discounted since the removal of traces of ions, such as sulphide or sulphate, at vanishingly low concentrations by ion-exchangers is very inefficient in the presence of a preponderance of chloride ions.

On consideration of method (iii), the use of an inert diaphragm with a pore size sufficiently small to prevent transfer of sulphur-containing ions from anolyte to catholyte would also limit rapid bulk transfer of electrolyte. Consequently, under the influence of the electric field, the catholyte would lose chloride ions through the diaphragm into the anolyte compartment, while the ferrous ion concentration would quickly decrease as the iron plated out on the cathodes. The catholyte conductivity would drop and eventually all electrolytic action cease. This state of affairs, however, could be prevented if the diaphragm consisted of a cation exchange resin membrane. Such a membrane would allow ferrous ions (Fe^{++}) to travel readily from anolyte to catholyte, but should prevent the transfer of chloride ions (Cl^-) from catholyte to anolyte, and simultaneously prevent the passage of sulphide (S^{--}) and sulphate (SO_4^{--}) ions in either direction. In practice, however, such a membrane is not 100% efficient and imperfections do allow slight transfer of anions across it. Nevertheless, under the conditions contemplated, although the voltage drop across the membrane would assist the transfer of anions (Cl^- in particular) from catholyte to anolyte, it should effectively bar their transfer in the opposite direction, thus preventing contamination of the catholyte with sulphur containing ions originating from the anode material.

Experiments with cation exchange resin membrane

Some preliminary work using cells divided by membranes of Permaplex C–10 (a commercial strongly acid cation exchange membrane supplied by Permutit Ltd, since superseded) gave the following evidence:

(i) water is transported through the membrane in conjunction with the ferrous ions, and that according to the particular batch of membrane used, 8–20 molecules of water accompanied each ferrous ion

(ii) some chloride ions penetrate the membrane in the direction catholyte–anolyte, the quantity varying from membrane to membrane

(iii) no detectable transfer of sulphate ions occurred from anolyte compartment to catholyte, even when the anolyte was heavily loaded with sulphate ions.

Both process (i) and (ii) lead to a reduction in the total catholyte concentration and an increase in the anolyte concentrations, and, in order to counter these effects, hydrochloric acid must be added to the catholyte at the same rate as chloride ions migrate through the membrane, while water must be added to the anolyte to make up that lost by transfer to the other compartment with the ferrous ions. The tendency for the concentration of the anolyte to increase may be countered either by the removal of the excess chloride ions or by continuous dilution of the anolyte with water, allowing the excess solution to run to waste via an overflow. Unfortunately the former method does not seem capable of a practical economic solution, although it could be achieved by electro-diffusing the chloride ions through an anion exchange resin membrane to be discharged at a carbon or iridium secondary anode. The simpler solution is to increase the rate of addition of water to the anolyte to include that required for the maintenance of its concentration at a constant value, as already suggested, and allowing the excess ferrous chloride to run to waste.

Considerable trouble was also encountered in the early stages of the investigation due to 'treeing', or the growth of dendrites. If this was not checked at an early stage, 'trees' grew through the membrane to the anode, short circuited the cell, and ruined the membrane.

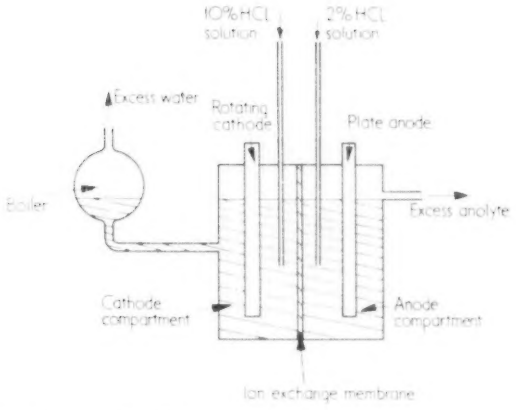
Various hypotheses have been proposed for the formation of these 'trees', but a sufficient reason in the present case seems to be the development of concentration cells in the system cathode–catholyte. These have the effect that the deposition of iron occurs more rapidly in solutions of higher concentration. Since higher concentrations tend to occur at the bottom of the cell, due to the higher density of more concentrated solutions, treeing is particularly pronounced at the bottom edges of cathodes. Also, when high current densities are used, the concentration of the solution in contact with the electrolyte is less than that in the bulk of the solution, so that any excrescence or edge tends to find itself in solutions of higher concentration than a plane surface. Thus, there is a tendency for plane, stationary electrodes in inadequately mixed solutions to grow trees at their edges, and also at any excrescence, as is observed experimentally. Furthermore, deposition tends to occur along surfaces in contact with the cathode, such as the surface of a hydrogen bubble attached to the electrode, thus generating excrescences on an initially plane surface.

Concentration cell effects can, of course, be cured by the addition of complexing reagents such as cyanide, oxalate, or tartrate, to reduce the ferrous ion concentration throughout the entire solution to a constant low value and maintain this concentration at the surface of the cathode by gradual dissociation of the complex. These effects can also be countered by vigorously stirring the solution. To minimize the risk of contamination, stirring was favoured in the present case.

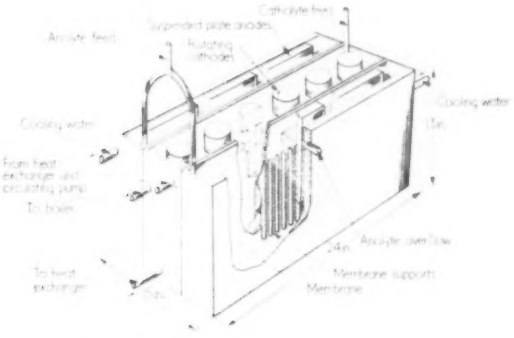
To avoid the effects caused by contact of hydrogen bubbles with the cathode surface, it was necessary to sweep these away as they formed. The simplest way of achieving this was to move the cathodes relative to the solution, and rotation of rod-shaped cathodes was found most suitable.

Pilot scale production

The investigations described above justified further work on a pilot scale. This was carried out in a cell constructed from a 2-V glass accumulator case split into two halves, with a C–10 membrane, sandwiched between perforated perspex sheets, clamped between these two halves to make a two-compartment cell. The anode compartment was 11 cm×3 cm×17 cm deep, while the cathode compartment was 11 cm×3·7 cm×17 cm deep, with an exposed area of the ion exchange resin membrane of 69 cm^2. The electrolyte concentration was maintained at 80–100 g Fe.l/in each compartment and at pH=2, while the whole cell was



1 *Experimental cell*



2 *Prototype cell*

immersed in a tank of running water to keep the temperature between 30 and 35 C. The catholyte was stirred vigorously by an immersed centrifugal pump and it contained two 1·25-cm dia. rods of pure iron as cathodes 80 cm long (immersed length) rotated at about 200 rev/min.

The equipment is shown diagrammatically in Fig.1. Considering first the anode compartment, 2%HCl solution (v/v) was added at the rate of 1 l./day to make up the water lost by transfer through the membrane, and to maintain the ferrous chloride concentration at a constant value. About 0·5 l. of ferrous chloride solution per day ran to waste.

10%HCl solution (v/v) was fed into the catholyte at a rate of about 0·6 l./day. The excess water introduced into the catholyte through the membrane and from the hydrochloric acid feed was removed by circulating the catholyte through a 200-W electric boiler which distilled off about 1 l. of water per day. The boiler was controlled by a constant level switch in the cathode compartment so that the value of the catholyte was maintained at a constant value.

This cell was run for 167·5 h with a current of 5 A at 5 V. The deposit was stripped every 24 h, and had the sulphur contents shown in the table below.

Deposit no.	S, %
1	0·0016
2	<0·0005
3	<0·0005
4	<0·0005
5	<0·0005
6	<0·0005
7	<0·0005
(probable error ±0·0002%)	
Sulphur content of anode iron	0·013%
Ampere efficiency	70%
Total mass of iron deposited	720 g

The figures indicate the evident success of the method in producing iron of the requisite quality. It was now necessary to scale the apparatus up to produce the requisite quantity, and accordingly a prototype equipment was designed that, it was hoped, would run semi-automatically, with a capacity of the order of 10 kg/week.

Prototype cell

The general layout and dimensions of the cell, which is made entirely of perspex, are shown in Fig.2. It contains 12 rotating cathode pins, each 10 in. long and tapering from $\frac{5}{8}$ in. dia. at the top to $\frac{3}{8}$ in. dia. at the bottom, suspended vertically at 2 in. centres. In order to accommodate the maximum effective area of membrane, the cathode compartment has an anode compartment placed on either side with two rectangular membranes, supported by perforated perspex spacers, separating the respective compartments. Each anode compartment contains six slab-shaped anodes of iron, each 9 in. × 4 in. × $\frac{1}{2}$ in. thick. The electrolyte is, as before, ferrous chloride solution at pH = 2 containing 100 g Fe/l., prepared by dissolving 5 kg of Norwegian electrolytic iron of low sulphur content (0·002%S) in high-purity hydrochloric acid ('Judactan' grade reagent, SO_4^{--} = 0·000075%). Thorough mixing of the catholyte is effected by withdrawing electrolyte from the base of the compartment and reinjecting it just below the surface of the solution. The electrolyte is circulated by a specially designed perspex centrifugal pump running in Teflon bearings, and while in the external circuit it is cooled by passing through a water-cooled heat-exchanger.

The liquid level in the cathode compartment is kept constant by slowly circulating the solution through a boiler in which excess water is evaporated off. The heater for the boiler is controlled by a float switch situated in the cathode compartment. Circulation of electrolyte through the boiler is necessary to avoid undue concentration of ferrous chloride building up in the boiler itself.

A metering force-pump supplies the requisite amount of 10% (v/v) hydrochloric acid from a reservoir to the catholyte to compensate for the depletion in chloride ion concentration due to transfer through the membrane. A similar pump supplies 2%HCl solution (v/v) to each of the anolyte compartments. The anolyte solutions are constantly stirred by a stream of nitrogen bubbles and excess ferrous chloride solution is allowed to run to waste from suitable overflows. To assist in removal of heat from the anolyte, water flows through coolant chambers forming the outer faces of the anode compartments.

The optimum conditions for the laying down of smooth coherent deposits could only be obtained by experiment, since they are dependent to a large extent on cell parameters. It was found that with the cell run

at 8–9 V, 70–85 A, and with the cathodes rotated at 100 rev/min, a fairly smooth, brittle deposit built up, giving a yield of 550–650 g of iron during a run of 7 h.

Under these conditions it was necessary to supply about 370 ml of 10%HCl/h to the catholyte, and 220 ml of 2%HCl/h to each of the anolyte compartments. The maximum cell temperature reached was below 30°C.

The cathode pins are fabricated from copper rod and to facilitate the removal of the deposited iron a layer of lead, 0·25 mm thick, is first electroplated on to the copper. It was at first envisaged that the electrodes would be heated to 400°C, whereupon the lead sheath would melt and the outer iron deposit drop off. However, it was found that the iron deposit was brittle enough to be knocked off the electrodes by smartly tapping with a light hammer. Residual lead is removed by pickling in a 75% hydrogen peroxide (20 vol.) 25% glacial acetic acid mixture.

During the course of operations a more robust membrane material became available (Permaplex C–20), but with similar ion-exchange properties, and it was this material that was used for the bulk of the operations. To date over 10 kg of iron have been deposited and analyses of typical 500 g batches are given in the table below, together with analysis of the anode iron.

The analyses indicate that the sulphur content of the deposited iron smoothly decreased as the catholyte became 'cleaned up' and soon fell to a value below the limit of detection of the analytical method used, which is described in the appendix.

Batch no.	C–10 membrane %S	Batch no.	C–20 membrane, %S
2	0·0022	10	0·0006
5	0·0018	16	0·0005
7	0·0013	21	<0·0005
9	0·0006	22	<0·0005
Anode iron	0·0095		

CONCLUSION

From the analytical results the sulphur content of the deposited iron has been very much reduced below that of the starting material. The content of other non-metallic impurities is similarly reduced, a typical analysis showing phosphorus reduced from 0·003% to 0·0004% and silicon from 0·008% to 0·003%.

That this method could be extended to the elimination of metallic impurities is not unreasonable, since many metallic ions can be converted to anions by the action of complexing agents, particularly organic reagents of the 'complexone' type.

ACKNOWLEDGMENT

The work described above has been carried out as part of the General Research Programme of the National Physical Laboratory, and this paper is published by permission of the Director of the Laboratory.

APPENDIX

Analytical method for sulphur

Principle

A large weight of sample is dissolved in a strongly oxidizing solution to convert the sulphur compounds to sulphate. The bulk of the iron is removed by a diethyl ether extraction, and the sulphur precipitated and determined as barium sulphate.

Special reagents

Hydrochloric acid (sp. gr. 1·10) Dilute 1·5 l. of hydrochloric acid (sp. gr. 1·16) to 2·4 l. with water.

Barium chloride solution (10% w/v) Dissolve 100 g of barium chloride in 1 l. of hydrochloric acid solution (2% v/v).

Acid wash solution Dilute 20 ml of hydrochloric acid (sp. gr. 1·16) to 1 l. The hydrochloric acid should be selected for particularly low sulphur content. The determination should be carried out in a clean, sulphur-free atmosphere.

Procedure

Transfer 12·5 g of sample to each of four 1·5 l. squat beakers and add 0·1 g of potassium bromide to each. Cover the beakers with clock-glasses. Add to each beaker 40 ml of nitric acid (sp. gr. 1·42) and then 60 ml of hydrochloric acid (sp. gr. 1·16) in increments of 5 ml to 10 ml, swirling the contents before each addition.

Digest until solution is complete. Remove the clock-glasses and wash both them and the sides of the beakers carefully. Evaporate the solutions to dryness and bake the residue on a hotplate to expel all traces of nitric acid. Add 75 ml of hydrochloric acid (sp. gr. 1·16) to each residue and heat until dissolved. Evaporate the solutions to a thick syrup.

Transfer the contents of each beaker to individual 500 ml separating funnels with 50 ml of hydrochloric acid (sp. gr. 1·10) and rinse each beaker twice with 25 ml of hydrochloric acid (sp. gr. 1·10) adding the rinsings to the main solution. Add 200 ml of diethyl ether to each funnel and shake for 2 min to extract the iron. Cool the separating funnels under a running tap during the extraction and release the pressure periodically.

Run the aqueous phases into separate 250 ml beakers and heat on a steam-bath to remove the residual ether. Evaporate to a syrup on a hot plate and then combine the four volumes in a single separating funnel. Use 100 ml of hydrochloric acid (sp. gr. 1·10) to effect this transfer. Extract a second time with 200 ml of diethyl ether as before, transfer the aqueous phase to a 250 ml beaker, drive off the residual ether and evaporate to a syrup. Take up the residue with 100 ml of dilute hydrochloric acid (10% v/v) and filter to obtain a clear solution. Add 10 ml of barium chloride solution (10% w/v) to the filtrate, cover, and digest on a steam-bath for 1 h. Remove from the bath, and set aside overnight.

Filter the solution through a tightly packed paper pulp pad and wash with the acid wash solution until the filter is free of iron salts. Then wash several times with cold distilled water until the washings give no reaction for chloride ion. Ignite the filter pad and precipitate at a low temperature until free of carbonaceous matter and then heat strongly at 850°C to constant weight. Cool and weigh the residue of barium sulphate.

A simultaneous blank determination on all reagents should be carried through all the stages of the process, and its weight deducted from the test weight.

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1. B. E. Hopkins *et al.*: *JISI*, 1951, **168**, 377–383.
2. G. A. Moore: *J. Met.*, 1953, **5**, 1443–1449.
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Factors influencing the adherence of oxides on metals

R. F. Tylecote

SYNOPSIS

An attempt has been made to explain the difference in the adherence of oxide films on iron, nickel, chromium, and cobalt by differences in the physical properties of the films. The ultimate tensile strength of thin specimens of NiO, α-Fe_2O_3, *and wüstite have been measured at elevated temperatures. The strength of* NiO *was found to reach a maximum of 9 300 lb/in², at 900°C and that of* α-Fe_2O_3 *5 000 lb/in², at 1 100°C. The UTS of wüstite was much lower than the other oxides. It was sensitive to loading rate, and creep effects were observed. In no case was a measurable elongation attained.*

The relative expansion coefficients of metal and oxide were compared. The negligible difference occurring in the case of nickel and nickel oxide was thought to explain the extremely good adherence of films on nickel after cooling to room temperature. 1827

INTRODUCTION

A REVIEW of the literature[1] indicated that more information was required on the mechanical properties of the oxides at elevated temperatures, particularly their capacity for deformation; the relative expansion coefficients of metal and oxide, and the effect of the compressive stresses in the base of the oxide film.

Before these points were investigated in detail, a general investigation of the adherence of oxides on various metals was made to determine the difference in behaviour.

Except where stated, all tests were carried out by the insertion of a specimen into a hot furnace thermostatically controlled at the temperature of the experiment. The furnace was mounted vertically, the specimens being hung on the end of a platinum/platinum–rhodium thermocouple; air being admitted by convection and oxygen, when used, being passed through at the rate of 20 l./h. After oxidation, the specimens were withdrawn from the furnace and allowed to cool in still air at room temperature.

ADHERENCE OF FILMS

Nickel

The composition of the high purity nickel strip used is given in column *A*, Table I. Oxidation between 860° and 900°C, resulted in completely adherent films, even when the oxidized metal was bent after oxidation.

Oxidation between 950° and 1420°C, produced thicker films which were also completely adherent. Bending after oxidation resulted in cracking of these films, but the oxide on the tension side remained adherent in spite of this treatment.

It appears that nickel oxide films on this grade of nickel are adherent on cooling after oxidation at all temperatures. This leads to the conclusion that there is something peculiar to the oxidation of nickel which results in such marked adherence. Figure 1*a* shows the fracturing of the oxide after bending, and demonstrates its extreme adherence. Closer examination of the interface (*see* Fig.1*b*) shows a rough interface with grain-boundary penetration of oxide, which is much more marked than on other metals of similar purity. This suggests that the intergranular penetration causes 'keying-on' of the scale, although pure copper (Fig.1*c*) shows a completely smooth interface with no grain-boundary penetration, and gives good film adherence. This is probably because of the extreme plasticity of its oxide at high temperatures.[2]

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Cobalt

The purity of the cobalt metal available was not high (about 99·0% excluding oxygen, see Table I, column *D*) and adherence was very poor, fracture occurring during cooling through a layer of fine-grained oxide near the interface (Fig.2). This area probably contains oxides of the impurity elements, and if so good adherence could not be expected under such conditions. The temperature range investigated was 500–1110°C.

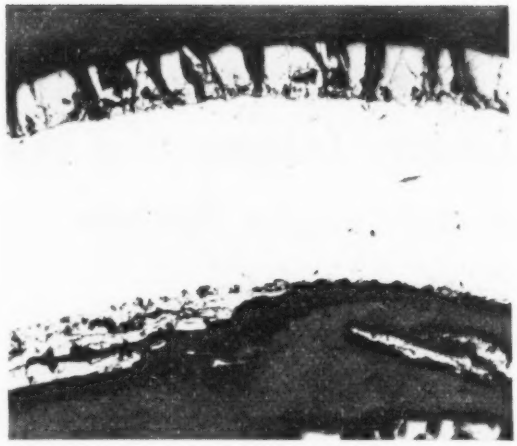
Chromium

Electrolytic chromium (>99·9%Cr) was oxidized in air at 735–1175°C. No spalling occurred on cooling from between 735° and 880°C. But spalling did occur after oxidation at 900°C for 18 h, and generally after shorter times at higher temperatures. A photomicrograph of the scale formed on chromium after 18·5 h at 890°C (Fig.3), shows a thin adherent blistered scale with a metal consumption zone showing intense porosity. It would appear that the scale on chromium has a high inherent strength, and that thin films adhere well. Thick scales, however, lose contact due to the enormously reduced area of metal at the base of the film.

The result of this test leaves little doubt that the oxidation of chromium occurs by the outward movement of chromium ions.

TABLE I Composition of metals investigated

Element	*A* Nickel sheet, %	*B* BISRA Iron sheet ABO, %	*C* JM specpure iron	*D* Cobalt roundels, %
Si	0·02	0·011	2 ppm	...
Fe	0·06	about 99·5	...	0·35
Cu	0·02	0·014	1 ppm	0·15
Mn	0·01	0·15	3 ppm	...
Mg	0·01	...	1 ppm	...
Cr	0·01	0·0036	...	...
Ti	Nil	0·003	...	...
Al	Nil	0·002	...	...
Ni	99·87	0·006	2 ppm	traces
C		0·03	0·03%	...
S		0·009	...	0·05
P		0·020	...	...
O_2			0·01%	about 2·0
N_2			0·01%	...
CaO			...	0·07



1a *Nickel strip oxidized for 4½ h at 1420 C and bent cold* ×70

Iron

Oxidation of iron containing 0·15%Mn (for composition *see* Table I, column *B*) failed to give adherent films in either air or oxygen in the range 500–1200 C. Although obvious spalling of scale did not occur on cooling in all cases, subsequent microexamination always showed a gap between the scale and the metal. Smooth interfaces were obtained with no grain-boundary penetration.

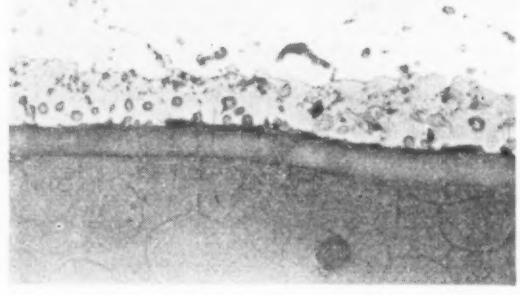
Microexamination revealed extremely variable amounts of the three oxides, showing that blistering was taking place during oxidation, very often causing the complete loss of contact of the scale on one side of the specimen.

The use of Remco iron containing 0·02%Mn, 0·04%C, and 0·005%S, caused an apparent improvement, but microexamination again revealed a gap between the metal and scale. From the evidence of scale composition it appeared that in this case the gap only formed upon cooling, and resulted from differential contraction.

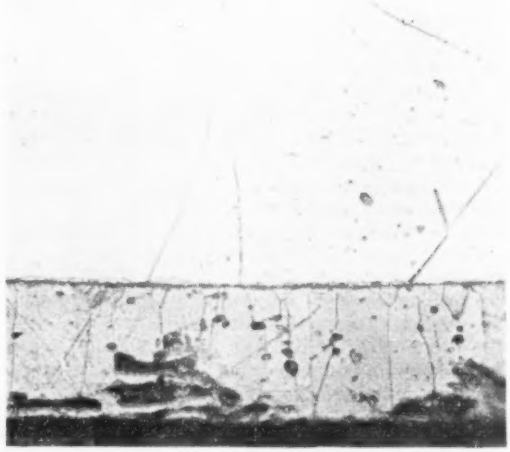
The precautions suggested by Païdassi[3] such as electropolishing and heating-up to oxidation temperature in an inert atmosphere, produced little improvement in the degree of adherence.

It was only upon the use of Johnson Matthey 'Specpure' iron (Table I, column *C*) that a great improvement was noted, and adherent films could be obtained upon iron in the range 600–1000 C. Little difference in adherence was noted between heating up the specimens in an inert atmosphere and introducing them directly into a hot furnace containing air. Some difference in adherence was found as between abrasive micropolished, rolled, and electropolished surfaces, but the main conclusion with respect to iron is that purity is the most important factor in obtaining adherent films.

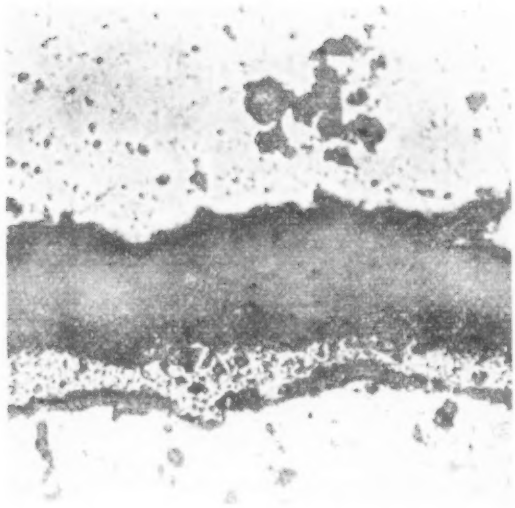
Peters and Engell[4] have shown that impurities which form oxides with higher free energies of formation than iron itself tend, as would be expected, to reduce the adherence of scale on iron. In this respect both 0·16%Si and 0·1%Al produced a considerable reduction in adherence, but the effect of 0·15%Mn was not so marked. In fact, the manganese content had to be increased to 0·63% before a reduction in adherence



1b *Nickel strip oxidized for 4½ h at 1420 C showing intergranular penetration of oxide* ×80



1c *Copper oxidized for 1 h at 900 C showing smooth interface and good adherence* ×150

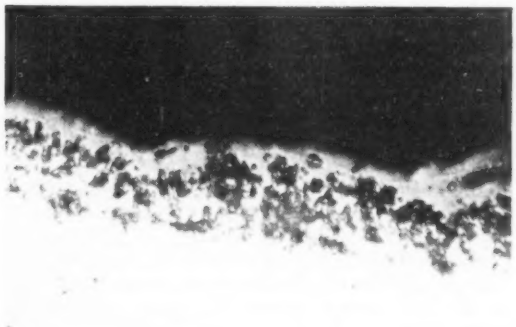


2 *Cobalt after oxidation at 1020 C for 24 h showing fine-grained porous oxide layer of CoO, probably containing oxides of impurity elements* ×300

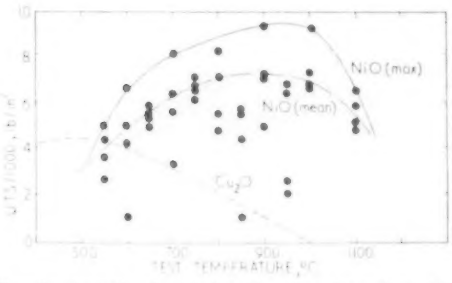
of the same order as that obtained with 0·16%Si was achieved.

Increase in the level of the more noble impurities such as copper, sulphur, phosphorus, and carbon, also produced a decrease in adherence. Therefore, we must conclude that the increase in adherence found in the present work with increasing purity is more likely to be connected with the general level of purity than any specific element.

Perhaps the main point that arises from these introductory experiments is the extraordinary adherence of the film on nickel. Although the purity is of the same order as the low manganese Remco iron sheet (about 99·88%Ni) the adherence is very different. This suggests that there must be either a great difference in the mechanical properties of the oxides, or the 'keying-on' effect must be important. In the case of chromium it would appear that the high-temperature properties of both metal and oxide were responsible for the maintenance of contact during oxidation and during cooling. In spite of considerable porosity at the base of the scale due to metal consumption, the scale has remained attached, probably due to the high metal strength at elevated temperatures.



3 *Electrolytic chromium after oxidation for 18½ h at 890 C* ×660



4 *Tensile strength of nickel oxide wires formed at 1150 C, Cu_2O included for comparison*

STRENGTH OF METAL OXIDES AT ELEVATED TEMPERATURES

Previous work on cuprous oxide has shown that it has a strength of 1–2 tons/in², an elongation of up to 20% in the range 400–800 C.[2] Such properties were thought to be sufficient to explain the relative adherence of such films on copper in spite of a large difference in the coefficient of expansion of metal and oxide.

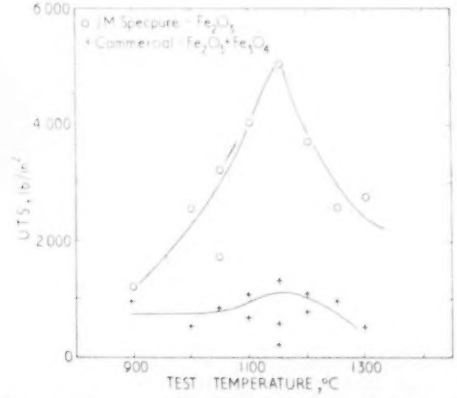
It was decided to determine the strength of nickel and iron oxides at high temperatures, since the former oxide is readily adherent upon nickel, and the latter oxides are generally non-adherent.

Nickel oxide

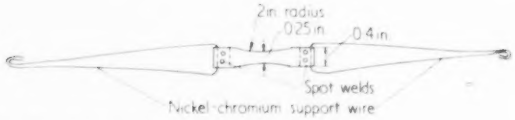
High purity nickel wire of 0·010 in. dia. was oxidized in air at 1150 C for 24 h. The wire, which hung vertically, was kept straight by a 10 g weight. No increase in length was found to have occurred during oxidation. After oxidation, an apparently solid wire of nickel oxide (NiO) was formed with a 0·0125 in. dia. giving an apparent density of 6·67 g/cm². Since the theoretical density of NiO is 7·45[3], this indicates a density of 90% of the theoretical. Microexamination showed that the porosity was evenly distributed across the section of the oxide wire.

The completely oxidized wires were loaded to fracture, at various temperatures, by pouring sand into a container weighing 6 oz over a period of 10 s.

The breaking loads at various temperatures are given in Fig.4. They show some scatter, but considering that the material appears to be fairly brittle at such temperatures, giving no measurable elongation or reduction in area in any specimen, scatter would be expected with such rapid loading.



5 *Strength of iron oxide wires formed by oxidation in air at 1050 C*



6 *Tensile test specimens used for iron strip*

The breaking stresses are very much higher than those obtained with copper (Fig.4). The decrease with lowered temperature is no doubt connected with the increased brittleness of oxides at low temperatures.

Iron oxides

Wires, 0·031 in. dia., made from commercial iron and 'Specpure' iron were oxidized at 1050°C in air, and tested over a range of temperatures. The specimens from 'Specpure' material were all found to be non-magnetic and single-phase, and by X-ray diffraction analysis were proved to be pure α-Fe_2O_3. Specimens made from the commercial material were mainly Fe_2O_3, but all were slightly magnetic and contained some Fe_3O_4, and a trace of FeO.

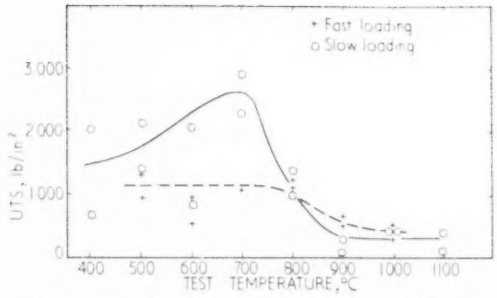
Metallographic examination showed that the wires had not oxidized concentrically. Oxidation of pieces of sheet showed how this had occurred. After a certain time the growing oxides break away from the metal on one side, producing a cavity. As a result the oxidation rate is reduced on this side and the oxide formed converted to Fe_2O_3. Growth continues on the other side until all the metal has been consumed. This effect has recently been reported by Engell and Wever.[6]

In the case of the single-phase specimens from 'Specpure' iron, there was the usual dense outer layer, and a porous inner layer on the side which had been in contact with the metal to the last. Calculations based on a specific gravity of 5·24 for Fe_2O_3 showed that the porosity of the oxide derived from 'Specpure' iron was 38% and from the impure 25%. In the results, which are given in Fig.5, allowance has been made for this porosity and the values for ultimate tensile strength relate to the true areas of oxide. The elongation in all cases was negligible.

The apparent reduction in strength of the pure hematite in the lower range of temperatures is probably connected with the extreme brittleness of this material. Even so, if it be a real effect, its strength at 900°C is still appreciably greater than that of wüstite, shown below. The strength of the oxide made for the commercial purity material shows little change with temperature. Its comparative weakness is probably due to the concentration of impurities in the grain boundaries.

A further series of experiments was carried out in an atmosphere of nitrogen saturated with water vapour at 90°C. In this atmosphere, wüstite is the stable phase, and oxidation rates are still sufficiently high to convert 0·005-in. thick iron sheet to oxide in 25 min at 1000°C.

The specimens consisted of strips 0·4 in. wide reduced to a width of 0·25 in. in the centre. The iron strip was looped round nickel–chromium support wires (*see* Fig.6) and at first spot welded. Later, it was found that the spot welding was unnecessary as the two thicknesses bonded together when completely oxidized.



7 *Effect of rate of loading on the strength of wüstite formed from 0·02%Mn iron at 1000°C*

Initially the rate of loading was the same as with NiO and Fe_2O_3, but later the effect of slow loading over periods up to 5 h was investigated, as it was apparent that the rate of loading was an important variable.

The slow loading experiments were carried out using a lever-loading technique in which a weight was moved along a lever in slow stages. Incremental loads of 0·2 lb, amounting to about 60 lb/in², were applied at intervals of 5 min. In many cases failure occurred as much as 3 min after the last increment had been applied, suggesting viscous creep. Yet in no case was a measurable elongation obtained.

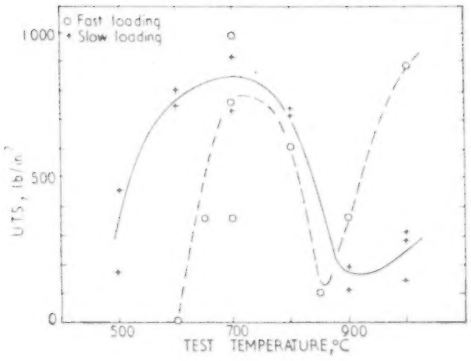
Figure 7 shows the results of fast and slow loading tests on wüstite made from high purity iron containing 0·02%Mn. It may be seen that the strength under fast loading conditions is far lower than that of the high purity Fe_2O_3, shown in Fig.5.

Under slow loading rates an appreciable increase in strength is obtained at intermediate temperatures. The maximum strength obtained is still half that of high purity Fe_2O_3. The fact that the two curves in Fig.7 cross, indicates that there is an appreciable creep effect at high temperatures. The higher strength with slower loading rates at lower temperatures is probably due to the inherent brittleness of oxides at lower temperatures and their high shock sensitivity.

Figure 8 gives the results for wüstite formed from the lower purity iron containing 0·15%Mn. The relative position of the fast and slow loading curves is the same as in Fig.7 and the same effects are therefore present. In this case, however, the effect of creep at high temperatures is more pronounced, resulting in a more marked reduction in strength under slow loading conditions. At lower temperatures the difference is not as great as in the case of the purer oxide.

Summary of results on tensile strength of oxides

The results of the tensile tests have been summarized in Table II, and for comparison the previously reported results for Cu_2O are included. The high strength of hematite compared with that of wüstite is clearly brought out at high temperatures. NiO has the greatest high-temperature strength of all the oxides investigated. The effect of slow loading is to increase the apparent strength of wüstite at temperatures between 400° and 700°C, but there is no such effect at higher temperatures. Delayed failure after an incremental load increase shows that creep is occurring. Recent work by Mackenzie and Birchenall[7] showed pronounced creep in wüstite at 900°C. Plastic deformation was also noticed at 800°C under slow bending.



8 *Effect of rate of loading on the strength of wüstite formed from 0·15%Mn iron at 1000°C*

However, no creep effects were noticed in Fe_3O_4 and Fe_2O_3, which they concluded, were stronger than wüstite. Recent Soviet work[8] showed that wüstite is far weaker than either Fe_3O_4 or Fe_2O_3. But the tensile test results are of quite a different order from those obtained in the present work, and it seems that the type of specimen used might be responsible for the difference.

It may be concluded that wüstite has some degree of plasticity under slow stressing which might account for its adherence under favourable conditions. But this plasticity is far less than that found for cuprous oxide in previous work where a considerable degree of elongation was obtained. The strength of nickel oxide is very much higher than any of the iron oxides, or cuprous oxide, but no sign of extreme ductility has been found, although the possibility that it has a ductility of the same order as wüstite cannot be dismissed.

The plasticity of wüstite is relatively unaffected by the presence of impurities, since creep effects were found in wüstite formed from both the 0·02%Mn and 0·15%Mn-containing iron. This applies only to the bulk material and not necessarily to the material at the metal/oxide interface of an oxidizing specimen where some degree of segregation could be expected to occur. During the oxidation of metal specimens to completion, impurities segregating at the metal/oxide interface would be moved towards the centre of the final specimen and form an impure core which would have little effect on the tensile properties. Only those impurities that occur in the grains themselves or the grain boundaries of the bulk oxide could be expected to exert a large influence on the tensile properties. This certainly happens in the case of phosphorus in copper. The differences between Figs.7 and 8 must also be explained by such a mechanism. Here, manganese has reduced the strength of the oxide at low temperatures, probably by causing embrittlement, but it has also increased the fast loading strength at high temperatures.

RELATIVE THERMAL EXPANSION OF METAL AND OXIDE

Since many metal oxides have defect sites in the metal or oxygen lattice, which may change concentration with oxygen pressure and temperature, it is necessary to pre-treat them to render them as stable as possible, and to specify this treatment. For the same reason the atmosphere in which the expansion is measured is also stated.

A Gale-type dilatometer, as used for phase changes in metals, was used. This is enclosed and permits very limited access of air during the measurements. The metal specimens measured 1·0 in. long by 0·25 in. dia., but some of the oxide specimens were shorter than this due to pressing and sintering difficulties.

Pressure and sintering conditions of the oxides are given in detail in Table III. The figures for chromium oxide have been supplied by Lucas[9] who carried out his measurements in this department using similar equipment, but much lower compacting pressures. As will be seen, the density of the sintered compacts exceeded 74% of the theoretical in all other cases.

No permanent change in dimensions resulted from the dilatometric tests, except in the case of the copper oxides. In these cases changes in shape occurred due to testing in the plastic range. However, in neither case was there any change in volume, and the specific gravity was found after test to be the same as before test.

TABLE II Results of tensile tests

Oxide	Test temperature, °C	UTS, lb/in²	Rate of loading
Cu_2O	500	4500	Fast
Cu_2O	900	1000	
α-Fe_2O_3	900	1000	Fast
α-Fe_2O_3	1150	5000	
FeO (0·02%Mn)	500	1100	
FeO (0·02%Mn)	700	1100	Fast
FeO (0·02%Mn)	1000	400	
FeO (0·02%Mn)	500	1700	
FeO (0·02%Mn)	700	2600	Slow
FeO (0·02%Mn)	1000	400	
NiO	500	3000	
NiO	700	8000	Fast
NiO	900	9300	
NiO	1100	6500	

SINTERING CONDITIONS AND RESULTS

The results obtained are given in Table IV, but some of the effects noted during the tests deserve comment.

Copper oxides

In the case of cuprous oxide, between about 650° and 900°C, rapid contraction takes place due to softening which continues on cooling, down to 510°C, where it ceases. Upon reheating the specimen from room temperature, the coefficient of expansion was increased to about 5×10^{-6}, and softening delayed until 800°C was reached. The specimen was still cuprous oxide, but it would appear that the increased coefficient was due to the change in the concentration of Cu^+ vacant sites upon heating in the small amount of oxygen present in the dilatometer, since after reheating in nitrogen at 900°C, the original coefficient was regained.

The coefficient of expansion of CuO was found to be $9{\cdot}3 \times 10^{-6}$ up to 600°C. At about 615°C, contraction occurred due to softening, showing that the oxide has some plasticity in this range. At about 850°C, loss of weight occurs, probably due to dissociation according to the equation:

$$4\ CuO \rightarrow 2\ Cu_2O + O_2$$

TABLE III Sintering conditions of oxide expansion specimens

Type and supplier	Compressing pressure, tons/in²	Sintering conditions: Temperature, °C	Time	Atmosphere	Theoretical specific gravity, %
NiO (BDH)	24·3	1400	20 h	air	83·5
Cu_2O (BDH)	24·3	800	2 h	N_2	85·0
CuO (Lab. reagent grade)	24·3	900	4 days	air	74·6
Fe_2O_3 (99%)	24·3	1100	6 h	air	91·0
CoO (Lab. reagent grade)	24·3	1200	16 h	N_2	75·4
Cr_2O_3 (Specpure)	2·2	1100	100 h	air	...

Cobalt

Cobalt metal undergoes a phase change at about 450° C, causing a contraction on cooling equivalent to about 12% of the total change in length between 450° and 20° C. The coefficient between 20° and 350° C was found to be $18{\cdot}6\times10^{-6}$, somewhat larger than the normally accepted figure over the range 25–350° C, of $14{\cdot}0\times10^{-6}$.[13] The coefficient in the range 800–900° C was $16{\cdot}9\times10^{-6}$. The impurities present in the cobalt used (Table I, column *D*), may have been responsible for these differences.

Iron

The phase change from α to γ iron upon heating in the neighbourhood of 900° C results in a contraction and an increase in the coefficient in the γ range. This contraction is of the order of 1% of the volume and might have a marked effect on the adherence of scales formed above 910° C.

Discussion

The results of the thermal expansion experiments are shown in Table IV, together with some recent coefficients obtained from published sources. The difference between the figures for metal and oxide, are an indication of the state of stress in the film when it is formed at temperatures below any phase change in the metal.

The difference in the coefficients for iron and wüstite would cause a difference in length of 0·3% on cooling from 900° C. However, upon cooling from 1000° C the phase change at 910° C would result in a linear expansion in the metal of 0·37%. This would subject the scale to a large tensile strain which might lead to it becoming non-adherent unless it possessed extreme plasticity. If the film is still adherent after this phase change the stress on it would gradually decrease, until at room temperature it would be subject to a slight tension. A larger difference would be obtained in the case of cobalt due to the phase change at 450° C. Nickel and NiO on the other hand, would give a very small difference of the order of 0·05% on cooling from 1000° C.

TABLE V Estimated compressive stresses in oxide films due to differential thermal contraction on cooling

Metal (form present at oxidation temperature)	Oxide	Cooling temperature range, °C	Compressive stress in film, tons/in²
α-Fe	FeO	900	3·63
γ-Fe	FeO	1000	−0·65
β-Co	CoO	900	1·04
Ni	NiO	1000	0
Cr	Cr_2O_3	1000	2·82
Cu	Cu_2O	900	17·3

Cuprous oxide and copper show the greatest difference, 1·14% when cooled from 800° C, while that for CuO and copper is 0·75%. Since the oxides are reasonably plastic in the upper part of the range, only some of the stress resulting from the difference in the coefficients will be actually applied to the scale on cooling.

The difference in the coefficients of chromium and Cr_2O_3 would cause a difference in length, on cooling from 1000° C, of the order of 0·22%. This is high, yet the scale is reasonably adherent, which suggests that the peculiar porosity of the metal, noted in the previous section, coupled with a certain degree of ductility might be the cause of its comparatively good adherence.

The compression stresses in these films may be calculated assuming $E=1{\cdot}3\times10^3$ tons/in², which has been suggested by Dankov and Churaev[14] as being representative of metal oxides. By plotting the expansion curves for oxide and metal on the same graph, it is possible to find the difference in length after cooling from a chosen upper temperature to room temperature. Due to the fact that the expansion characteristic for metals tends to be a curve and that for an oxide a straight line, the difference in length at room temperature tends to be more than proportionally greater as the oxidation temperature is increased. However, it is in the high-temperature portion of the characteristic that plastic deformation is most likely to occur, and this will therefore tend to reduce the internal stress by a large amount.

The figures given in Table V have been calculated on the assumption that there is no plastic deformation; with the exception of copper they are in accordance with the known facts on adherence. The large stress expected for Cu_2O is not achieved, due to plastic deformation. The absence of stress in NiO, would assist a strong brittle oxide film to adhere, while the small stress in chromic oxide is probably relieved by deformation of the porous metal.

TABLE IV Linear expansion coefficients of metals and oxides

System	Oxide Temperature range, °C	Coefficient, $\times10^{-6}$	Reference	Metal Temperature range, °C	Coefficient, $\times10^{-6}$	Reference
Fe/FeO	100–1000	12·2	Rigby[11]	0–900	15·3	Hidnert[12]
Fe/Fe_2O_3	20–900	14·9*				
Ni/NiO	20–1000	17·1*		0–1000	17·6	ICT[10]
Co/CoO	20–900	15·0*		25–350	14·0	Metals Handbook[13]
Co/Co_3O_4	20–900	13·6	Lucas[9]	600–900	15·9	Metals Handbook[13]
Cr/Cr_2O_3	100–1000	7·3	Rigby[11]	0–1000	9·5	Lucas[9]
Cu/Cu_2O	20–750	4·3*		0–800	18·6	Metals Handbook[13]
Cu/CuO	20–600	9·3*				

* Results of present work; rest from published sources, in which the copper and iron used were electrolytic, and the nickel purity was 99·06%

CONCLUSIONS

As far as iron is concerned, the evidence obtained in this investigation shows that under certain conditions there should be a marked difference in the adherence on cooling between specimens cooled from an oxidation temperature of 900 C and those cooled from 1000 C. If the plasticity of the scale formed at 1000 C is sufficient for it to remain intact on cooling below the γ–α transformation temperature, the tensile stress in the film caused by transformation will be progressively reduced, with only a small tensile stress remaining at room temperature. Furthermore, if there is a residual compressive stress in the film arising from the oxidation process, the dimensional changes caused by the transformation process in the iron would be expected to reduce it. Thus, given some degree of plasticity and a slow cooling rate, scales formed at 1000 C might be expected to be more adherent than those formed at 900 C, provided they were of the same thickness. In the absence of large amounts of the higher oxides, a scale consisting predominantly of wüstite should have some plasticity which may offset any tendency to spall during oxidation. However, this limited degree of plasticity can be expected to be relatively ineffective in preventing spalling on rapid cooling.

Nickel has a very strong oxide film with marked evidence of 'keying-on', and there is no difference between the expansion coefficient of the metal and that of the oxide. All these factors go towards explaining the adherence of the film both during oxidation and cooling after oxidation.

The cobalt tested was relatively impure, and this fact probably contributes towards the poor adherence. The low temperature transformation from β to α would tend to reduce the stress in the film, but in the presence of the complex oxide film between the more massive scale and the metal, the slight stress remaining is sufficient to break contact between metal and scale.

Chromium shows an extraordinary high degree of porosity in the metal at the base of the film, and it is clear that further work is required on this metal. The difference in the expansion coefficients of metal and oxide is relatively great, and from this point of view good adherence would not be expected.

The expansion data for cuprous oxide shows the importance of plasticity. The copper-cuprous oxide combination has the greatest difference in expansion coefficient and forms a smooth interface with little keying-on. But there is no sign of a porous layer near the interface, and abundant evidence of great plasticity in the film. The result is relatively good adherence of high-temperature scales, although those formed at lower temperatures, with a high proportion of the less plastic CuO, do not show such good adherence.

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The kinetics of the martensite transformation in iron alloys

W. S. Owen and A. Gilbert

SYNOPSIS

Martensite transformations in iron alloys are classified according to the crystal structure of the product. The driving force required for the transformation in Fe–Ni *alloys is compared with that necessary in* Fe–Cr *and* Fe–C *alloys. The kinetics of martensite formation is discussed briefly in terms of the nucleation of new martensite plates and the variation with temperature of the fraction of austenite transformed. Some recent ideas about stabilization effects are reviewed.* 1885

A DETAILED ACCOUNT of the kinetics of martensite reactions was published only two years ago by Kaufman and Cohen.[1] The historical development of the subject, Krisement's adiabatic hypothesis, the early reaction-path theory, and the attempts to apply homogeneous nucleation theory are adequately discussed. As far as is possible, this material will not be duplicated. This short review is confined to iron-base alloys. In spite of these self-imposed limitations, sufficient new information has been published since the Kaufman and Cohen review to warrant further general discussion and it seems particularly important to direct attention to those areas in which at present there is a surfeit of conflicting ideas.

It is first necessary to define a martensitic transformation. Originally, attention was focused on the unusual properties of the microconstituent produced in steel on quenching, but this view is much too narrow to serve as a basis for a definition. Most recent definitions[1, 2, 36] have stated dual criteria, (*a*) that all atom displacements are less than one atomic spacing, so that no diffusion in the normal sense is involved and (*b*) these displacements are such as to produce a lattice shear which is manifest by a tilt produced on a plane free surface. Bilby and Christian[36] consider the shear displacement to be the essential feature, the diffusionless nature of the transformation being merely a corollary. Since in all cases the composition of the parent phase and the transformation product is identical, the first criterion appears to be general. However, as will be discussed later, it is not always possible to demonstrate experimentally that the transformation in iron alloys involves a lattice shear. It might be argued that such reactions should not be classed as martensitic, but this view is not adopted here.

A necessary, but not always sufficient, condition for a transformation to occur is that the total Gibbs free-energy of the system be reduced. On slow cooling the decrease in free energy is achieved by the initial phases transforming to phases of different composition. For example, at the eutectoid temperature the free energy of an entirely austenitic iron–carbon alloy is equal to that of the same specimen transformed to ferrite and cementite of equilibrium composition. If austenite is retained to a lower temperature the free energy of the systems is greater than that of a pearlitic specimen and a driving force exists, equal to the difference in the total free-energies, tending to transform the austenite to pearlite. However, such transformations require the transfer and rearrangement of atoms over large lattice distances and they can occur only if sufficient time is allowed for the diffusion processes. Martensitic reactions, being diffusionless, derive the driving force from the difference in free-energy between the initial and final phase, the two phases differing only in crystal structure. This implies that there is a temperature T_0 at which the free energies are equal. The first problem in any kinetic study is to find the value of T_0 and to determine the driving force at the reaction temperature. Very many diffusionless transformations in iron alloys are athermal. That is, the start temperature is independent of the cooling rate and the extent of the transformation is determined solely by the temperature to which the specimen is cooled. Most theoretical treatments are concerned with the temperature at which martensite first forms, M_s, and consequently the measurement of this temperature has been the subject of much experimental work. However, since Kurdjumov and Maximova[15] first reported the isothermal formation of martensite in a steel containing 0·6 wt-%C and 6·0 wt-%Mn many other instances of isothermal martensite formation have been reported. For example, martensite can form isothermally in 16–8 and 18–8 stainless steels[8] and in iron–nickel–manganese alloys.[15] The temperature must be below T_0, but can be either above or below M_s, as determined for the athermal transformation. Metallographically isothermal and athermal martensites are indistinguishable.

Since martensite forms at many different locations and subsequently increases in volume at the expense of the austenite, the overall transformation is conveniently considered in terms of two distinct stages, nucleation and growth. The growth of martensite plates or lenticular particles occurs extremely rapidly in iron alloys. Bunshah and Mehl[3] measured growth

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rates of 10^5 cm sec^{-1} in an iron–29·5%Ni alloy. Thus, it is usual to consider that the activation energy for growth is effectively zero and that the kinetic problem is essentially one of nucleation. This is true of both athermal and isothermal martensite formation. In the former case the number of nuclei is related directly to the temperature, but in the latter the number increases with time at the reaction temperature.

TYPES OF MARTENSITE IN IRON ALLOYS

It is convenient to classify martensite transformations in iron alloys according to the product.

1. Tetragonal martensite is formed by quenching alloys containing carbon. The axial ratio, c/a, is independent of the austenitizing temperature and the cooling rate and, according to Kurdjumov[2]

$$c/a = 1 + 0{\cdot}0467 \times \text{wt-\%C}$$

It has been reported that iron–nitrogen martensite is also tetragonal[4] and it is tempting to generalize from this that all martensite formed in interstitial alloys is tetragonal but, due to experimental difficulties, evidence from other interstitial alloys is sparse.

2. Cubic martensite, indistinguishable from ferrite of the same composition is formed by quenching iron–nickel,[5–7] iron–manganese,[5] iron–chromium,[11,12] and other binary[12] and ternary alloys.[8–10,15,41]

3. An hexagonal constituent is formed on quenching Fe–Mn alloys containing 13–25%Mn[13,14] and with Ni, Cr, Co, and C additions,[16] Cr–C steel[22,23] and Fe–Ni–Cr alloys.[8,16] This phenomenon is intimately connected with the formation and generation of hexagonal stacking faults and manganese and chromium increase the susceptibility to faulting to a much greater extent than does nickel.[10] The hexagonal constituent often occurs associated with cubic martensite. Otte[10] considers that it is unlikely that martensite is nucleated by stacking faults, but Venables[60] has produced evidence which suggests that in an 18–8 stainless steel the hexagonal constituent is a step on the way to cubic martensite formation.

With alloys in which, under equilibrium conditions, there is an extensive $(\alpha+\gamma)$ phase field at a temperature corresponding to the M_s, it is a simple matter to demonstrate that on rapid quenching no chemical segregation occurs and so the first criterion is satisfied. However, when considering low nickel or manganese alloys and those systems in which a γ-loop forms, other more indirect evidence of chemical homogeneity has to be adduced. Provided a cooling rate faster than a critical rate is used, the transformation start-temperature cannot be depressed by further increasing the rate.[17,18] With Fe–Ni alloys containing 5 at-%Ni the critical cooling-rate is slower than 5°C/min but with Fe–Cr alloys a rate of at least 1000°C/s is required and with a 2·7 at-%Si alloy the critical rate is probably about 5000°C/s.[12] The transformation start temperatures recorded at faster than critical rates vary smoothly with alloy content, but, unlike the transformation temperatures for undercooled diffusion reactions, do not follow the equilibrium phase-boundary profile. Further, these start temperatures extrapolated smoothly from those determined for high-alloy specimens in which the temperature of transformation is much lower than that at which a significant number of diffusion jumps could be expected.

Only the tetragonal martensites and the more highly alloyed cubic martensites satisfy unambiguously the second criterion. Hexagonal martensite produces markings on pre-polished surfaces, but it has not been demonstrated that these are due to a homogeneous shear. In low-alloy cubic martensite, including Fe–Ni,[6] the surface eruptions produced by the transformation are not well defined, nor can plane interfaces between martensite and retained austenite be distinguished in polished and etched specimens.[5,6,12,19]

It seems improbable that long-range diffusion is involved in the polymorphic transformations in pure iron under any conditions, since the transfer of atoms across an interface will occur more rapidly than diffusion of identical atoms from within the matrix. In this sense all transformations in pure metals are diffusionless. They satisfy the first, but not the second, criterion. However, there have been a number of reports that a martensitic reaction, satisfying both criteria, can be induced in iron if very rapid quenching is employed.[20,21,12]

The three types of martensite have other common characteristics. In all those cases in which decomposition by a diffusion process can be avoided by rapid heating the martensites transform to the original austenite at a temperature A_s above T_0. In the case of Fe–Ni alloys it has been demonstrated that T_0 is halfway between M_s and A_s.[7] Further, all the transformations are assisted by plastic deformation of the austenite. The martensite starts to form at a higher temperature (M_d) on cooling and the reverse transformation starts at a lower temperature on heating. These aspects of the subject have been fully discussed in recent reviews.[1,36]

THE DRIVING FORCE

To start a martensitic transformation an alloy has to be undercooled from T_0 to M_s because energy has to be available to supply the non-chemical energies, such as interfacial energy and strain energy. The available energy at any temperature T is the difference between the free-energy of austenite, F^γ, and the martensite, $F^{\alpha'}$, at that temperature. This difference, usually called the driving force, is

$$\Delta F^{\alpha'\to\gamma} = F^\gamma - F^{\alpha'} \quad \ldots\ldots(1)$$

and is positive at temperatures at which martensite is the more stable phase. Using Kaufman and Cohen's[1] notation, for a binary alloy Fe–A, where A is carbon, nickel, chromium, manganese, or silicon, etc., the free energy of the austenite is

$$F^\gamma = (1-x)\,F_{Fe}{}^\gamma + xF_A{}^\gamma + F_M{}^\gamma \quad \ldots\ldots(2)$$

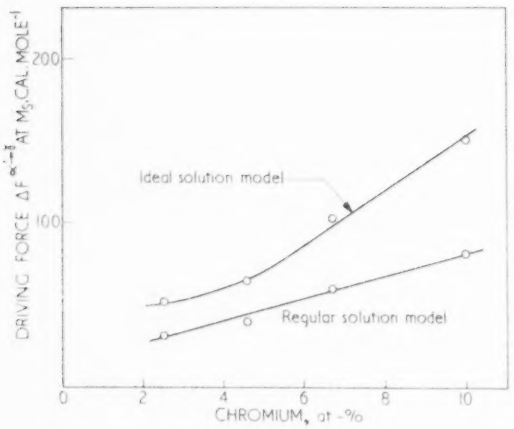
x is the atomic fraction of element A and $(1-x)$ the fraction of iron. $F_{Fe}{}^\gamma$, $F_A{}^\gamma$, and $F_M{}^\gamma$ are respectively the free energy of pure iron, pure element A and of mixing A and Fe, all in the austenitic state. Similarly, in the martensitic state.

$$F^{\alpha'} = (1-x)F_{Fe}{}^\alpha + xF_A{}^\alpha + F_M{}^{\alpha'} \quad \ldots\ldots(3)$$

Since the first two terms on the right-hand side of the equation refer to pure elements in the ferritic state α' is replaced by α. Subtracting equation (3) from equation (2)

$$\Delta F^{\alpha'\to\gamma} = (1-x)\Delta F_{Fe}{}^{\alpha\to\gamma} + x\Delta F_A{}^{\alpha\to\gamma} + \Delta F_M{}^{\alpha'\to\gamma} \quad \ldots\ldots(4)$$

Johansson[24] and Zener[25] calculated $\Delta F_{Fe}{}^{\alpha\to\gamma}$ from specific heat measurements collected by Austin.[26] The matter has been further discussed by Fisher[27] and by



1 *Driving force $\Delta F^{\alpha'\to\gamma}$ at M_s as a function of chromium content in* Fe–Cr *alloys*[12]

Kaufman and Cohen[7] who proposed the following relation

$$\Delta F_{Fe}^{\alpha\to\gamma} = 1202 - 2{\cdot}63 \times 10^{-2}T^2 + 1{\cdot}54 \times 10^{-6}T^3 \quad \ldots\ldots(5)$$

This expression gives values which are in fair agreement with the Fisher[27] interpolated values, it satisfies the Nernst condition that $\partial(\Delta F)/\partial T = 0$ at 0 K and it has the virtue that it is moderately accurate over a wide temperature range (200–1000 K). Since the term involving $\Delta F_{Fe}^{\alpha\to\gamma}$ is much larger than the other terms in equation (4) it is necessary that it be evaluated with the best possible accuracy. For the purpose of discussing M_s temperatures in low-alloy cubic martensite the range 800–1100 K is of particular interest and the present authors[12] have suggested a modified expression

$$\Delta F_{Fe}^{\alpha\to\gamma} = 1474 - 3{\cdot}4 \times 10^{-2}T^2 + 2 \times 10^{-6}T^3 \quad \ldots\ldots(6)$$

which also satisfies the Nernst conditions and gives a closer fit than the Kaufman and Cohen[7] expression in this restricted temperature range.

The second term of equation (4), $x\Delta F_A^{\alpha\to\gamma}$, usually refers to one or more hypothetical modifications of the element A and so its value cannot be obtained by direct experiment. The free energy of mixing term, $\Delta F_M^{\alpha\to\gamma}$, could be obtained experimentally from activity measurements of element A in iron in the austenitic and martensitic state, but suitable measurements have not been reported. Thus, it is necessary to attempt to evaluate these two terms from thermodynamic models of the solid solutions. Fisher[27] adopted a model which led to the result that the mixing term

$$\Delta F_M^{\alpha\to\gamma} = RT \ln f_A^{\gamma}/f_A^{\alpha} \quad \ldots\ldots(7)$$

where f_A^{γ} and f_A^{α} are the activity coefficients of element A in austenite and martensite of the same composition. Making the further assumption that the term $x\Delta F_A^{\alpha\to\gamma}$ is negligible, he obtained for the driving force

$$\Delta F^{\alpha'\to\gamma} = x_{Fe}\Delta F_{Fe}^{\alpha\to\gamma} + x_c(-10{\cdot}500 + 3{\cdot}425\ T) + \Delta F^* + \Sigma x_A RT \ln f_A^{\gamma}/f_A^{\alpha} \quad \ldots\ldots(8)$$

x_{Fe}, x_c, and x_A are the atom fractions of iron, carbon, and element A respectively. The term ΔF^* takes account of the decrease in free energy which accompanies the spontaneous ordering of carbon when body-centred tetragonal martensite is formed. Cohen, *et al.*[28] also derived an expression for the driving force for carbon martensite which agrees well with Fishers. Surprisingly, it appears that for carbon martensite $\Delta F^{\alpha\to\gamma}$ at the M_s temperature is independent of the carbon content (and therefore of the M_s temperature). The value obtained is 290 cal mole^{-1}.

Zener[25] assumed that low-alloy steels can be treated as dilute solutions so that the ideal solution model is applicable. By definition of an ideal solution $\Delta F_M^{\alpha'\to\gamma}$ is zero and

$$\Delta F_A^{\alpha\to\gamma} = RT \ln x_{\alpha}/x_{\gamma}$$

where x_{α} and x_{γ} are the compositions at the α and γ phase boundaries at temperature T. In the Zener model the simplifying assumption is made that $RT \ln x_{\alpha}x_{\gamma}$ is constant, which is tantamount to assuming that the entropy change $\Delta S^{\alpha\to\gamma}$ is also zero. Then

$$\Delta H_A^{\alpha\to\gamma} = RT \ln x_{\alpha}/x_{\gamma} \quad \ldots\ldots(9)$$

where $\Delta H_A^{\alpha\to\gamma}$ is the heat of transformation in the pure element A. Equation (4) becomes

$$\Delta F^{\alpha'\to\gamma} = (1-x)\Delta F_{Fe}^{\alpha\to\gamma} + x\Delta H_A^{\alpha\to\gamma} \quad \ldots\ldots(10)$$

Jones and Pumphrey[5] applied equation (10) to transformations in Fe–Ni alloys, but since $\Delta H_A^{\alpha\to\gamma}$ for this system is clearly not constant, serious difficulties were encountered.[1] In view of this, Kaufman and Cohen,[7] have attempted to evaluate each term of equation (4) using a regular, instead of an ideal, solution model. From the known equilibrium boundaries for temperatures between 573 and 1183 K, F_M^{α} was obtained and it was possible to write an explicit expression for the driving force as a function of temperature and nickel content. However, this expression could be evaluated only by neglecting a term which becomes significant at greater than 25 at-%Ni. To extend the analysis to higher nickel contents, values of T_0 from experimental data (taken to be $\frac{1}{2}(M_s + A_s)$) had to be used. The final expression is given in ref.1. This analysis can be applied only to the Fe–Ni system in which certain difference-terms are sufficiently small to be negligible.

Systems with a γ-loop present special difficulties. As an example the Fe–Cr case will be discussed. If the ideal solution model is adopted and the values of x_{α} and x_{γ} corresponding to the maximum of the γ-loop are substituted in equation (10) a value of $\Delta H_{Cr}^{\alpha\to\gamma} = 150$ cal mole^{-1} is obtained. At T_0 $\Delta F^{\alpha'\to\gamma}$ is zero and

$$(1-x)\Delta F_{Fe}^{\alpha\to\gamma} = -x\Delta H_{Cr}^{\alpha\to\gamma} \quad \ldots\ldots(11)$$

Elements with positive values of $\Delta H^{\alpha\to\gamma}$ increase the stability of the α-phase. At a fixed composition, T_0 must lie between the equilibrium temperatures indicated by the boundaries of the two-phase field and since in the Fe–Cr case the $(\alpha+\gamma)$ field is narrow this defines T_0 within fairly close limits. With increasing chromium content T_0 first decreases, reaches a minimum and subsequently increases until the loop is closed at about 13%Cr. Thus, $\Delta H_{Cr}^{\alpha\to\gamma}$ would have to be negative at low chromium contents and positive between about 8 and 13%Cr. A similar difficulty arises when the effect of chromium on M_s is considered. Assuming that the effects of carbon and various substitutional elements on the M_s of alloy steels is additive, it has been estimated from experimental data that each wt-% of Cr *lowers* the M_s temperature by 19°C.[29] Zener[30] and others[1] have assumed that this could be taken to be the variation of M_s with chrom-

ium content in the binary Fe–Cr system in which cubic martensite is formed, although in the alloy steels examined the martensite was tetragonal. A recent determination[12] of M_s of the binary alloy as a function of chromium content has shown that the true depression of M_s is not very different: ΔM_s is −17 °C/1 wt-%Cr. Thus, between 8 and 13%Cr, increasing the chromium content increases T_0 but decreases M_s. As Zener[25] pointed out, this amounts to saying that chromium acts as an alpha stabilizer at high temperatures and a gamma stabilizer at low temperatures. Since $\Delta H_{Cr}^{\alpha\to\gamma}$ is assumed to be independent of temperature, the continued decrease of M_s with increase in chromium content beyond 8%Cr can only be explained in this model, if the driving force for martensite formation increases very markedly with increasing chromium content. The driving force estimated from the recent experimental results for M_s as a function of chromium content and using the value $\Delta H_{Cr}^{\alpha\to\gamma} = 150$ cal mole^{-1} is shown in Fig.1. $\Delta F^{\alpha\to\gamma}$ increases from 15 cal mole^{-1} at 1 at-%Cr to 80 cal mole^{-1} at 10 at-%Cr.

Zener[30] proposed a solution to the problem which depends upon splitting the free-energy change into a magnetic and a non-magnetic component. This leads to the expression,

$$\Delta F^{\alpha'\to\gamma} = \Delta F_{Fe}^{\alpha\to\gamma}(T - 20x_{Cr}) - 1{\cdot}41\,x_{Cr}(\Delta M_s - \Delta T_{mag}) \quad \ldots(12)$$

where ΔT_{mag} is the temperature shift produced by the magnetic uncoupling effect. Unfortunately, there appears to be no independent way of estimating ΔT_{mag} since its relation to any measurable quantity, such as the Curie temperature, is not known. Thus ΔT_{mag} can be treated only as a disposable parameter.

Application of the regular solution approximation to binary alloys with a restricted γ-field has been made possible by the work of Svechnikov and Lesnik[31] who, from a study of equilibrium relationships, obtained an expression

$$\Delta F_{Cr}^{\alpha\to\gamma} = 460 - 1{\cdot}0\,T \text{ cal mole}^{-1} \quad \ldots(13)$$

Only the free energy of mixing term is needed to complete equation (4) and this is given directly by the regular solution model

$$F_m^{\gamma} = Bx(1-x) + RT(x\ln x + (1-x)\ln(1-x)) \quad \ldots(14)$$

and

$$F_m^{\alpha} = Ax(1-x) + RT(x\ln x + (1-x)\ln(1-x)) \quad \ldots(15)$$

subtracting (15) from (14)

$$\Delta F_m^{\alpha\to\gamma} = (B-A)x(1-x) \quad \ldots(16)$$

Putting equations (16) and (13) into equation (4)

$$\Delta F^{\alpha'\to\gamma} = (1-x)\Delta F_{Fe}^{\alpha\to\gamma} + x(460 - 1{\cdot}0\,T) + (B-A)x(1-x) \quad (17)$$

To obtain a value for the empirical constant (B–A) Kaufman[32] considered the variation of T_0 (taken to be midway between the α and γ equilibrium boundaries) with chromium content. With $\Delta F^{\alpha'\to\gamma} = 0$ the best fit with the T_0 curve was obtained when

$$(B-A) = (-2800 + 0{\cdot}75T) \text{ cal mole}^{-1} \quad \ldots(18)$$

Thus, for the Fe–Cr system

$$\Delta F^{\alpha'\to\gamma} = (1-x)\Delta F_{Fe}^{\alpha\to\gamma} + x(460 - 1{\cdot}0\,T) + x(1-x)(-2800 + 0{\cdot}75\,T) \quad \ldots(19)$$

As Kaufman has pointed out,[32] the dilemma presented by the Zener treatment is now resolved. $\Delta H_{Cr}^{\alpha\to\gamma}$ in equation (10) is now replaced by the sum of a positive term for $\Delta F_{Cr}^{\alpha\to\gamma}$ and a negative term for $\Delta F_m^{\alpha\to\gamma}$. At high temperatures the positive term dominates, and chromium acts as a ferrite stabilizer, but at low temperatures the mixing term becomes more important, and the austenite is stabilized.

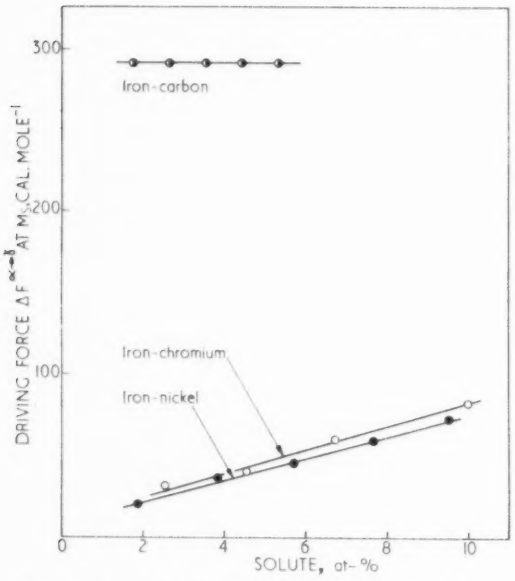
In Fig.1 the driving force for martensite formation in Fe–Cr alloys calculated on the basis of the ideal solution model, equation (10) is compared with that required by the regular solution model, equation (19). The values in the latter case are much smaller and the increase with chromium content is less marked.

When compared at the same atom fraction of solute, T_0 for Fe–Ni and Fe–Cr are markedly different but the driving force required to bring about the martensitic transformation at M_s is very nearly the same in the two cases (Fig.2). For Fe–C alloys the energy needed at M_s is at least six times larger, and there is no variation with carbon content. However, at high nickel content (30 at-%), the discrepancy between the Fe–Ni and Fe–C alloys disappears.[7] Large values for the driving force have been reported also for Fe–Cr–Ni alloys.[32] Thus, there seems to be a correlation between the metallographic appearance and the driving force; poorly defined structures formed in alloys with relatively low alloy content requiring only a small driving force compared with that required for the formation of well-defined lenticular plates of martensite in carbon or highly alloyed steels. However, until the reason for the increase of driving force with nickel or chromium content is understood, and much more experimental data on other systems are available, this suggestion cannot be more than speculation.

THE NUCLEATION OF MARTENSITE

In terms of the classical concept of nucleation[33, 37, 38] in condensed systems, the stability of a small volume of martensite (an embryo) in the parent austenite is determined by the relative magnitude of a number of energy factors. Assuming the embryo is a lenticular plate of radius r and thickness $2c$, these are:

1. *The driving force* This is a temperature dependent term equal to $\pi r^2 c\Delta f$. Where Δf is the driving force, $\Delta F^{\alpha\to\gamma}$ per unit volume of martensite.
2. *The energy of the interface* The surface energy σ varies little with temperature and the energy term is $2\pi r^2\sigma$. In the absence of an acceptable dislocation model for the interface it is difficult to estimate the value of σ with certainty. Using a model developed by Frank[34], Knapp and Dehlinger[35] have estimated σ for martensite in steel to be about 200 erg cm^{-2}.
3. *The elastic strain energy* It was assumed by Fisher and Turnbull[39] that the shear strain is confined to the lattice within a sphere of radius r. Then the strain energy is proportional to the volume of the martensite embryo and is $K\pi rc^2A$; K is a proportionality and A a strain energy constant. A is about $2{\cdot}1\times10^{10}$ erg cm^{-3} (500 cal cm^{-3}).[35, 1] Christian,[62, 64] and Kaufman,[63] have discussed this and the interfacial energy term in more detail than is possible here.
4. *Energy associated with dilatation strain within the embryo*
5. *An energy term resulting from any non-reversible plastic work done by the embryo on the austenite* In some transformations the accommodation strains are purely elastic, making possible the thermo-elastic effects, but in iron alloys it seems probable that the martensite plate can be accommodated only by slip occurring in the austenite matrix.



2 *Driving force $\Delta F^{\alpha'\rightarrow\gamma}$ at M_s for* Fe–C, Fe–Ni, *and* Fe–Cr *alloys*

6. *Kinetic energy* The last three terms are usually considered to be negligible. The energy change on the formation of an embryo is then written as

$$\Delta W = \pi r^2 c \Delta f + 2\pi r^2 \sigma + K\pi r c^2 A \quad \ldots\ldots(20)$$

At any value of T this expression is minimized with respect to r and c giving

$$c^* = -\frac{8\sigma}{3\Delta f}$$

$$r^* = -\frac{2Ac^*}{\Delta f} \quad \ldots\ldots(21)$$

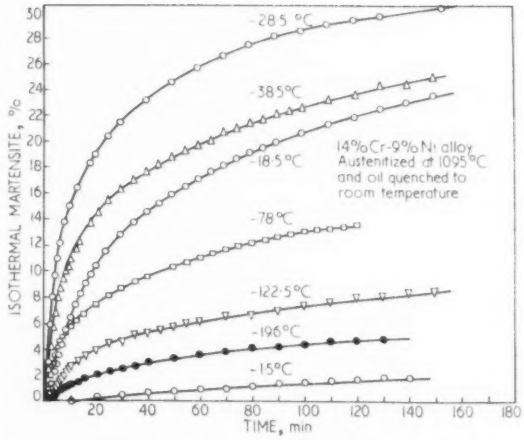
The corresponding value of the net energy change ΔW^* is the energy barrier to nucleation and embryos of critical size r^*, and c^* will grow spontaneously to full-size martensite plates with a decrease in free energy. Athermal martensite formation is easily explained on this basis if it is assumed that at each temperature there is a characteristic size distribution of embryos. As the temperature is decreased Δf changes sign on passing through T_0. At lower temperatures the critical size of embryo decreases with temperature and at each temperature below M_s the largest embryos present will become active nuclei.

Detailed models of martensite nucleation can be classified under three headings.

1. *Thermally activated nucleation*

The occurrence, under certain circumstances, of isothermal martensite strongly suggests the possibility that its nucleation is a thermally activated process. Fisher and Turnbull[39] have taken the view that this is the case under all circumstances, athermal martensite formation being merely a special case in which isothermal nucleation occurs very rapidly. Making the assumption that the nucleation is homogeneous, from standard chemical rate theory they obtain for the nucleation frequency:

$$I = N\frac{kT}{h}\exp\left(\frac{\Delta W^* + Q}{kT}\right) \quad \ldots\ldots(22)$$



3 *Rate of isothermal formation of martensite in an iron alloy containing 14%Cr and 9%Ni austenitized at 1095°C and oil quenched to room temperature (Kulin and Speich[43])*

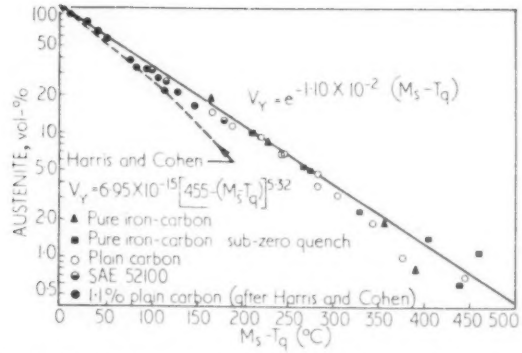
where N is the number of atoms per cm^3, k and h are the Boltzmann and Planck constants respectively, and Q is the activation energy to transfer an atom across the austenite/martensite interface. Since the rate of growth of martensite plates is extremely rapid, Q must be effectively zero. Because ΔW^* appears in an exponential term it dominates the expression and the nucleation frequency is very sensitive to changes in this saddle-point energy.

Fisher[40] applied classical theory with remarkable success to the experimental kinetic data obtained by Cech and Holloman[41] for isothermal martensite formation in an Fe–Ni–Mn alloy. However, Kaufman and Cohen[1] have demonstrated that this analysis is not of general application. The weakness of the theory is strikingly shown by an example quoted by Cohen.[42] Using the best available values for A and σ he calculated the size of the martensite nucleus for a 30 at-%Ni alloy at the athermal M_s temperature (233°K) and obtained $c = 22$ Å, $r = 490$ Å, and $W = 9 \times 10^{-9}$ ergs/nucleus. Such a nucleus would contain nearly two million atoms, and the large activation energy required to form it could not possibly be supplied thermally at 233°K.

The assumption of homogeneous nucleation is not essential to the classical thermally-activated nucleus hypothesis but the situation is not markedly improved by assuming heterogeneous nucleation.[38, 1]

2. *Strain nucleus theory*

These theories start from the assumption that martensite is nucleated at preferred sites. It is considered that the critical event is not reached when the nucleus is a critical size but that it is connected with the establishment of the conditions necessary to start some shearing mechanism. The heterogeneous nucleus is usually considered to be a volume of austenite lattice in which there is a local concentration of strain. The early theories of this type considered only displacements in the lattice[43] but recent models deal with metastable groupings of dislocations enclosing a volume in which the lattice approximates to martensite. It is necessary to specify that these embryos are already larger than



4 *Empirical relation between the extent of martensite formation and the temperature according to Koistinen and Marburger.*[52] *An equation due to Harris and Cohen*[57] *is also plotted*

the size required by classical nucleation theory when the critical condition for movement of the dislocation interface is reached.

The most detailed theory of this type is due to Knapp and Dehlinger.[35] Following Frank[34], they consider a martensite embryo with a $(225)_\gamma$ habit plane to be an oblate spheroid bounded by dislocation loops; growth in the $[1\bar{1}0]_\gamma$ and $[225]_\gamma$ directions being by expansion of the loops and in the $[55\bar{4}]_\gamma$ direction by the generation of new loops. When the driving force is sufficient to provide the energy to expand the interface and to supply the macroscopic elastic strain energy resulting from the formation of a martensite plate, sudden growth of a plate occurs. Thus the critical condition is visualized in terms of an energy balance similar to that set out in equation (20) with $\Delta W = 0$. The driving force needed to nucleate these embryos decreases with increasing embryo size. Originally, the model was applied only to the athermal case. As the temperature is decreased below T_0 the driving force increases and at M_s the largest embryos expand rapidly. At lower temperatures smaller embryos become critical and so the reaction continues. Kaufman and Cohen[1] have further elaborated the Knapp–Dehlinger model, and shown that it can be used to explain the kinetics of isothermal martensite formation in Fe–Ni–Mn alloys.

On this model it is anticipated that the driving force at M_s should be the same for Fe–Ni and Fe–Cr alloys because the transformation geometry, and therefore the dislocation arrangement in the interface, is the same in both cases and the strain terms will differ only by virtue of a small difference in the shear modulus. The results in Fig.2 substantiate this prediction. However, the much greater driving force required for the formation of Fe–C and the highly alloyed cubic martensites is difficult to explain on the basis of the Knapp–Dehlinger model. It may be that the difference lies in the strain term, but this must remain speculative until the martensite shear angles are measured as a function of alloy content.

The most striking aspect of this theory is the requirement that very large embryos exist in the austenite. For example, to explain the athermal and isothermal kinetics of a 23%Ni, 3·6%Mn alloy, it is necessary for embryos of radius 200 Å to be present above the M_s temperature. If such large heterogeneities exist in the austenite they can be studied by direct experiment. The results of the first investigation of this type have been reported by Richman *et al.*[44] Using thin-foil transmission electron microscopy they have observed 'plate-like markings' which it is claimed are martensite embryos. Unlike the classical case, the embryo size distribution is not a function of the temperature, since it is assumed that the embryos are frozen-in from the austenitizing treatment. Thus, the size at the temperature in the microscope is the same as at the M_s temperature. Measurements of the maximum size of the plates agrees reasonably well with that predicted for a number of binary alloys in the range 28·0 to 30·7 at-%Ni. However, a much more intensive study of the heterogeneities in austenite is required before the theory can be considered to rest on a substantial experimental base. In particular it needs to be shown that the 'plate-like markings' observed by transmission microscopy do in fact propagate to martensite plates at the M_s temperature.

3. *Dislocation generation models*

By the separation of suitable partial dislocations, one layer of austenite lattice can be transformed to martensite. To produce a plate of macroscopic thickness it is then necessary to devise a means by which the mechanism can be transferred to adjacent planes. One feasible way of doing this is by the 'pole-mechanism'[45] by which the rotating dislocation climbs on a helicoidal surface, and so transfers the transformation dislocation to successive planes. An alternative proposal involving the dynamic multiplication of dislocations has also been made.[46]

The most explicit theory of this type relevant to martensite formation in iron alloys is due to Jaswon.[47] It is assumed that the crystallographic relationship resulting from the transformation can be achieved by two shears; half the twinning shear along elements $\{111\}_\gamma\langle 112\rangle_\gamma$ and a second, minor, shear along $\{112\}_\gamma\langle 110\rangle_\gamma$. It is proposed that the first shear is produced by the dissociation

$$\frac{2d}{3}\langle 112\rangle = \frac{d}{3}\langle 112\rangle + \frac{d}{3}\langle 112\rangle \quad \ldots\ldots(23)$$

forming an extended dislocation with a stacking fault of martensite structure. At high temperatures this nucleus becomes very small, but below M_s the free-energy of the martensite is lower than that of austenite and the stacking fault expands very rapidly. Since it is difficult to understand how this expansion can be thermally activated, the theory in its present form is appropriate only to athermal martensite formation. As with the strain nucleus theories, it is implied that the critical event is the sudden release of the boundary. However, in this case it has not yet been demonstrated that the classical nucleation requirements, imposed by the fact that the interfacial area is increased and macroscopic elastic strains are produced, are automatically satisfied.

No way has been proposed of assessing the energy of the stacking fault of pseudo-martensite or the variation of this energy with temperature and alloy content. Thus, no quantitative deductions can be made from this model.

PROGRESS OF THE TRANSFORMATION

Once the martensite embryo has passed through the critical nucleating event, it propagates rapidly until it meets some obstacle. Radial growth is stopped at a

grain boundary, at the interface of another martensite plate, or possibly by some other macroscopic imperfection. The plates in any one specimen are surprisingly uniform in thickness and it is not clear why growth in the thickening direction should stop before the transformation is complete. It has been suggested that this is due to jamming of the interfacial dislocations[1] but a quantitative theory is lacking.

A typical isothermal transformation curve is shown in Fig.3. The most feasible explanation of the rapid decrease in reaction rate with time appears to be that the martensite plates partition the austenite so that new plates form in smaller austenite volumes and so transform less austenite. This model has been treated quantitatively by Fisher.[48]

When the martensite forms athermally the extent of the transformation is a function of the temperature only. A number of empirical relationships have been proposed relating the volume of austenite transformed to the temperature difference between M_s and the transformation temperature, T_q. Some recent data are shown in Fig.4.[52] These relationships are unaffected by variations in composition or in austenitizing conditions. Thus, once the M_s temperature is known, the volume fraction of austenite at any temperature can be predicted. Since the metallurgical variables affect the transformation through the effect on M_s, it is not surprising that thermodynamic treatments appropriate to the M_s can be applied without modification to describe the subsequent transformation. Brook, Entwisle, and Ibrahim[61] have examined the transformation curves for a number of Fe–Ni–C and Fe–Cr–C steels. They calculated $\Delta F^{\alpha\to\gamma}$ from Fisher's[27] equation (8) (substituting the Kaufman and Cohen[7, 32] values of $\Delta F^{\alpha\to\gamma}$ for Fe–Ni and the values for Fe–Cr from equation (19)) and showed that $\frac{\delta \Delta F^{\alpha\to\gamma}}{\delta T}$ is proportional to the slope of the transformation curve. Thus, the relationships found experimentally between the fraction transformed at any temperature T_q and the M_s temperature, arises from a fundamental correlation between the increment of transformation and the change in driving force with temperature.

If the cooling is interrupted or slowed down, martensite formation may be retarded significantly. This thermal stabilization effect can occur with both athermal[51] and isothermal martensite formation[41, 50] and the stabilizing temperature can be either above or below the M_s. When the reaction restarts it usually occurs with a measurable burst.[56] It has been stated that the martensite lost during the stabilization treatment is never recovered during subsequent transformation[51] but contradictory results have been obtained.[57]

Stabilization cannot be induced in steels which do not contain carbon or nitrogen.[8] This was clearly demonstrated by Philibert[54] who showed that all traces of a stabilization effect disappear from nickel steels when the carbon is removed by treatment in moist hydrogen. He determined the activation energy for the stabilization process as 11000 cal mole^{-1}. The effect was ascribed to hardening of the austenite by the segregation of carbon or nitrogen to dislocations. A similar suggestion has been made earlier by Morgan and Ko.[57] Crussard[58] claims that for stabilization to occur, deformation of the austenite is necessary, but that this is achieved easily either as a result of quenching stresses or the volume change accompanying the growing martensite plates. Edmondson and Ko[55] have shown metallographically that slip occurs in the austenite during the martensite transformation in nickel steels. Crussard[58] visualizes 'a kind of precipitation' occurring in the austenite, but there seems to be no evidence on which to decide between this proposal and the suggestion that the segregate is a Cottrell atmosphere. Woodilla *et al.*[59] have also determined the activation energy of the stabilizing process in a nickel steel and obtained the value 15500 ± 3000 cal mole^{-1}. They claim that this activation energy is incompatible with the view that carbon segregation in the austenite is rate controlling, and suggest that the essential diffusion process occurs in the martensite plates. Because the activity of the carbon in the martensite is much greater than in the austenite there is a driving force for diffusion towards the austenite. This produces segregation of carbon and nitrogen to the martensite embryo interface which thus becomes locked. Glover and Smith[56] also consider that it is diffusion of carbon and nitrogen within the martensite which leads to stabilization but they propose that first-stage tempering occurs, causing a contraction of the martensite, and so a relaxation of the elastic stresses in the surrounding austenite. In this way the favoured nucleation sites are destroyed, and the nucleation of new martensite plates is inhibited.

Clearly, further experimental evidence is required to permit a distinction to be made between the various theories of stabilization. The technique of thin-film electron microscopy should prove helpful in this connexion. Although stabilization is an important effect both practically and theoretically, it is easy to devise experiments to study martensite kinetics in which it does not intervene. Thus, it is not unreasonable to consider it to be a side effect.

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The mechanical properties of some tempered alloy martensites

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SYNOPSIS

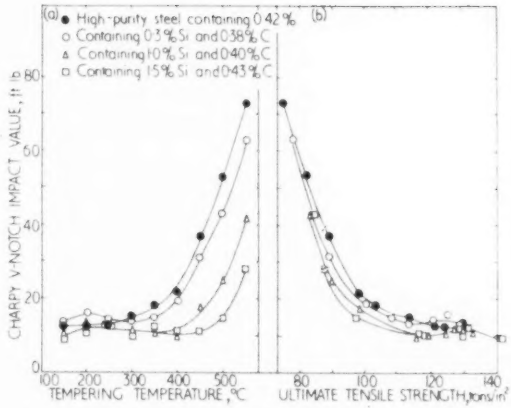
Tensile, impact, notched-tensile, fatigue, and notched-fatigue properties have been determined for high-purity, martensitic nickel–chromium–molybdenum–vanadium steels, heat-treated to high strength levels, and have been compared with the properties of similarly treated steels of commercial purity. It was found that impact properties, tensile ductility, notched-tensile strength, and notched-fatigue strength were improved by the removal of impurities, but plain-bar fatigue strength appeared to be unaffected by the purity of the steel. In both high-purity and commercial-purity steels, the addition of silicon (1·3–1·5%) slightly retarded the tempering of martensite; in high-purity steels it had a deleterious influence on impact properties and tensile ductility. In the absence of impurities, and of the associated '350°C-embrittlement' phenomenon, carbon content proved to have an outstanding influence on the impact properties of high-purity 1½%Ni–Cr–Mo steels. In such steels, lowering the carbon content raised impact values to levels considerably higher than those obtained with comparable steels of commercial purity. 1884

INTRODUCTION

ALTHOUGH hardened and tempered low-alloy steels are often described as martensitic, to differentiate them from other steels in which bainite or ferrite+pearlite form an important part of the as-quenched microstructures, direct-hardening steels are not usually put into service in the fully martensitic (untempered) condition. It is generally appreciated that optimum mechanical properties in hardened and tempered steels are obtained when *martensite* comprises the initial structure. It is, however, not always economical or practicable to employ steels which are essentially martensitic when hardened: e.g. it may give rise to problems such as distortion and cracking associated with the quenching of intricate shapes. Martensitic steels are used, therefore, where service conditions are particularly severe, and when maximum toughness is required at relatively high strength levels.

To attain high strengths, the tempering of martensitic steels is usually restricted to the range 100–500°C. After such tempering the acicular nature of the microstructure is retained to a degree which depends on the composition of the steel and the tempering temperature chosen: indeed, when the steels are only lightly tempered, it is difficult to differentiate, by normal optical microscopic examination, between the untempered and the tempered structures except that the latter are somewhat more reactive to etching reagents. The present paper is mainly concerned with the properties of a selection of martensitic steels in the lightly tempered condition. It is not possible, however, to define an upper tempering-temperature limit beyond which the structures can no longer be considered as martensitic, since this limit varies appreciably with variation in composition, and for certain steels may be quite high.

This investigation was initiated to explore the potentialities of low-alloy steels at high strength levels, i.e. greater than 100 tons/in², with particular reference to the specific requirements of the most recent aircraft designs and missiles. It is recognized that the trend in modern aircraft has introduced new problems in selection of materials, because at the high speeds already attained, and the still higher speeds that are envisaged, it is no longer a question of finding materials which have a high strength/weight ratio at normal atmospheric temperatures. As a result of aerodynamic heating, the requirement is for materials which will provide the requisite strength for a given weight at temperatures up to about 300°C, or, in the case of missiles, in an even higher range. At such temperatures, ultra-high-strength steels give strength/weight ratios superior to those of aluminium alloys,[1,2] but the use of steels of such high strengths introduces special problems. One of these is associated with the well known embrittlement phenomenon exhibited by most standard alloy steels when hardened and tempered in the range 250–400°C, a phenomenon sometimes referred to as '350°C-embrittlement',[3] and on account of this effect it has been the standard practice to avoid tempering in that range. The use of tempering temperatures below 250°C would, however, seriously curtail the usefulness of high-strength steels, since



a v. tempering temperature *b* v. UTS (from hardness figures)

1 *Influence of silicon on impact properties of high-purity* 1½%Ni–Cr–Mo *steels at room temperature*

after such treatment they would be structurally unstable at the (higher) operating temperatures of certain aircraft parts. If, on the other hand, the steels were tempered above 450°C, the resulting loss of tensile strength would preclude their use.

A similar problem has emerged in the re-designing of steel undercarriage components. Here again, the trend has been to up-grade the strength of low-alloy steels, and although operating temperatures may not be as high as in the case of wing- and fuselage-coverings, it is desirable to use steels which combine high strength with good ductility and toughness after tempering at temperatures above 250°C.

In recent years much information has been published on the mechanical properties of a wide range of alloy steels heat-treated to tensile strengths above 100 tons/in², and there are several examples of the application of such steels in aircraft.[4,5] Among the most important developments which have assisted progress in this field was the demonstration by Allten and Payson[6] that the addition of up to 2·18%Si to a 3%Ni steel retarded the tempering of martensite. Shih *et al.*[7] also showed that the addition of 1·5%Si to steels of SAE 4340 and 4325 types (1·8%Ni–Cr–Mo) retarded the softening of the martensitic structure on tempering, and made possible the use of somewhat higher tempering temperatures without introducing 350°C embrittlement. Silicon has subsequently been used as an alloying addition by other investigators[8,9] in the development of ultra-high tensile steels. More recently, Kenneford and Williams,[10] after a systematic examination of the influence of a number of elements on the hardness/tempering-temperature relationships for martensitic steels containing 0·35%C, concluded that chromium, molybdenum, and vanadium also had useful effects in retarding the onset of softening. However, they gave no information on the effects of these elements on impact properties.

In the various investigations concerned with development of ultra-high-strength steels, the principle has been accepted that, for optimum mechanical properties, the as-quenched structure should be completely martensitic, and the carbon content should not be higher than that necessary to give the requisite tensile strength. The work of Busby *et al.*[11] on the

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TABLE I Impurities in raw materials for the high-purity steels

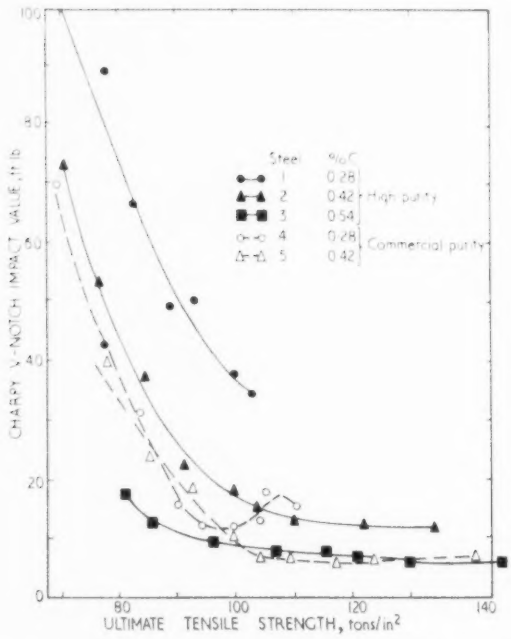
Material	Analysis, wt-% Al	Sb	As	Bi	B	C	Cu	Fe	Pb	Mg	Mn	Mo	Ni	P	Si	S	Sn	Ti	V
Carbonyl iron	*c.*0·007	ND	<0·01	...	ND	0·005–0·007	...	...	<0·002	<0·005	<0·01	<0·01	*c.*0·1	0·001–0·003	ND	<0·002	<0·01	ND	...
Nickel pellet	<0·005	ND	ND	ND	<0·005	0·018	0·002	0·019	<0·002	<0·005	<0·01	<0·01	...	...	0·002–0·005	0·001	<0·005	<0·01	<0·01
Electrolytic chromium	*c.*0·005	ND	ND	ND	ND	...	<0·01	0·2–0·5	0·0005	<0·005	0·02–0·05	*c.*0·01	...	...	<0·01	...	*c.*0·005	<0·01	<0·01
Molybdenum	0·05–0·1	...	...	...	<0·001	0·02–0·05	...	0·05–0·1	<0·003	<0·005	<0·02	...	...	...	0·02–0·05	0·005	...	...	...
Ferro-vanadium	...	<0·005	0·02–0·05	ND	<0·005	...	*c.*0·05	...	0·01–0·05	<0·01	*c.*0·5	<0·01	<0·01	...	*c.*1·0	...	0·02–0·05	<0·01	...

ND = Not detected

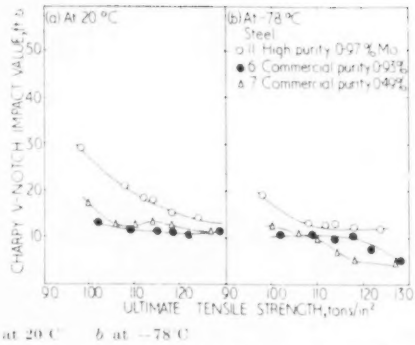
TABLE II Analyses of steels investigated

Steel no.	Purity	Type	Size of ingot, lb	Analysis, wt-% C	Si	Mn	Ni	Cr	Mo	V	S	P	Al	Ga	N	Pb	Sb	Ti	B
1	HP	1½%Ni–Cr–Mo	25 (V)	0·28	0·04	0·02	1·43	1·10	0·25	<0·01	0·002	0·003	0·015	ND	...	<0·002	*c.*0·001	<0·01	*c.*0·001
2	HP	1½%Ni–Cr–Mo	25 (V)	0·42	0·04	0·02	1·53	1·09	0·24	<0·01	0·003	0·001	0·045	ND	...	<0·002	*c.*0·001	<0·01	*c.*0·001
3	HP	1½%Ni–Cr–Mo	25 (V)	0·54	0·02	0·05	1·76	1·01	0·15	<0·01	0·002	0·001	0·010	ND	...	<0·002	*c.*0·001	<0·01	*c.*0·002
4	CP	1½%Ni–Cr–Mo	25 (A)	0·29	0·17	0·39	1·4	1·3	0·29	<0·01	0·028	0·008	0·025	ND	0·004	<0·002	*c.*0·001	<0·01	<0·001
5	CP	1½%Ni–Cr–Mo	25 (A)	0·42	<0·1	0·33	1·55	1·3	0·28	<0·01	0·029	0·009	0·010	ND	0·005	<0·002	*c.*0·001	<0·01	<0·001
6	CP	1·8%Ni–Cr–Mo–V	200 (A)	0·42	0·27	0·59	1·65	1·25	0·93	0·29	0·013	0·038	0·036	ND	0·004	<0·002	<0·001	<0·01	<0·001
7	CP	1·8%Ni–Cr–Mo–V (low Mo)	200 (A)	0·39	0·12	0·49	1·77	1·31	0·49	0·27	0·011	0·025	0·053	VFT	0·007	<0·002	*c.*0·001	<0·01	<0·001
8	CP	4¼%Ni–Cr–Mo–V	200 (A)	0·34	0·23	0·53	4·0	1·28	0·53	0·18	0·010	0·031	0·025	VFT	0·008	*c.*0·002	*c.*0·001	<0·01	<0·001
9	CP	4¼%Ni–Cr–Mo–V (high Si)	200 (A)	0·35	1·30	0·61	4·0	1·35	0·53	0·20	0·011	0·031	0·034	VFT	0·011	*c.*0·002	*c.*0·001	<0·01	<0·001
10	CP	1·8%Ni–Cr–Mo–V	25 (A)	0·42	0·43	0·59	1·8	1·20	1·05	0·28	0·014	0·007	0·043	VFT	0·004	<0·002	*c.*0·001	<0·01	<0·001
11	HP	1·8%Ni–Cr–Mo–V	25 (V)	0·43	<0·1	<0·1	2·0	1·30	0·97	0·27	0·001	0·001	0·028	VFT	<0·002	<0·002	*c.*0·001	<0·01	<0·001
12	HP	1·8%Ni–Cr–Mo–V	25 (V)	0·40	<0·1	<0·1	2·2	1·30	0·018	0·30	0·001	0·003	0·017	ND	<0·002	<0·002	*c.*0·001	<0·01	<0·001
13	HP	4¼%Ni–Cr–Mo–V	25 (V)	0·34	<0·1	<0·1	4·6	1·25	0·58	0·20	0·002	0·001	0·028	VFT	<0·002	<0·002	*c.*0·001	<0·01	<0·001
14	HP	4¼%Ni–Cr–Mo–V + 1·5%Si	25 (V)	0·36	<0·1	<0·1	4·6	1·40	0·49	0·23	0·001	0·003	0·016	ND	<0·002	<0·002	*c.*0·001	<0·01	<0·001
15	HP	4¼%Ni–Cr–Mo–V + 1·5%Si	25 (V)	0·34	1·45	<0·1	4·1	1·40	0·57	0·24	0·003	0·001	0·060	VFT	<0·002	<0·002	*c.*0·001	<0·01	<0·001
16	HP	4¼%Ni–Cr–Mo–V + 1·5%Si	25 (V)	0·36	1·45	<0·1	4·3	1·35	0·56	0·26	0·001	0·003	0·050	ND	<0·002	<0·002	*c.*0·001	0·046	<0·001

HP = High-purity V = Vacuum melt ND = Not detected
CP = Commercial-purity A = Air melt VFT = Very faint trace
Note: As, Co, Cu, Mg, Sn, and Zr were <0·01%; Bi and Ge were not detected.



2 *Influence of carbon content and purity on impact properties of hardened and tempered* $1\frac{1}{2}$%Ni–Cr–Mo *steels*


(a) At 20 °C
(b) At −78 °C
Steel
○ 11 High purity 0·97% Mo
● 6 Commercial purity 0·93%
△ 7 Commercial purity 0·49%
CHARPY V-NOTCH IMPACT VALUE, ft lb
ULTIMATE TENSILE STRENGTH, tons/in²

a at 20 °C *b* at −78 °C

3 *Room-temperature and sub-zero impact properties of commercial-purity and high-purity 1·8%Ni–Cr–Mo–V steels*

mechanical properties of low-carbon martensites lends considerable support to the view that the lower the carbon content, the better are the ductility and impact properties obtained for a given tensile strength. These investigators also showed that nickel was most beneficial in lowering impact-transition temperatures of very lightly tempered martensite, and that up to 2%Cr also improved impact properties to a limited extent. Manganese, on the other hand, in amounts ranging from 0·8% to 1·5% or more, raised transition temperatures.

Several claims have been made that melting and casting *in vacuo* improved the fatigue properties of high-strength alloy steels.[12,13] These improvements were most notable in the case of carbon–chromium ball-race steels, and it seems probable that a reduced inclusion content may have been an important contributing factor.

Rinebolt,[14] investigating the tensile and impact properties of steels of the SAE 4340 type melted in air, *in vacuo*, or in argon, and subsequently heat-treated to tensile strengths between 67 and 130 tons/in² found that no benefits were conferred by melting either *in vacuo* or in argon. On the other hand, one of the present authors[15] demonstrated that in vacuum-melted and -cast Ni–Cr–Mo steels made from pure base materials, '350°C-embrittlement' was eliminated. Tests on similarly prepared steels to which various impurity elements had been added indicated that such embrittlement could be induced by the presence of small amounts of phosphorus, arsenic, antimony, tin, and nitrogen, or by larger quantities of manganese (up to 2%) or silicon (up to 1·5%). The nitrogen content referred to (0·02%) was higher than would normally be encountered in commercial steels, but the results obtained suggested that vacuum melting could have beneficial effects on the impact properties of tempered martensite, provided that the contents of other impurities were also low.

The influence of silicon is of particular interest, in view of the practical use which has been made of this element in several types of ultra-high-strength steel developed in the USA for aircraft applications. Relationships between impact value, tempering temperatures, and tensile strength, for high-purity $1\frac{1}{2}$%Ni–Cr–Mo steels of varying silicon content, reproduced in Fig.1, indicate that with increasing silicon the tempering range in which embrittlement is encountered is raised towards 450°C.

The absence of '350°C-embrittlement' in high-purity Ni–Cr–Mo steels suggested the possibility of developing improved ultra-high-strength steels on this basis, since it was expected that higher impact values might be obtained at strength levels obtainable only by tempering at temperatures which, for commercial-purity steels, would be in close proximity to, if not within, the embrittling temperature range. Work was therefore undertaken to provide further information on the mechanical properties of high-purity martensites alloyed with nickel, chromium, molybdenum, vanadium, and silicon, after tempering to high strength levels.

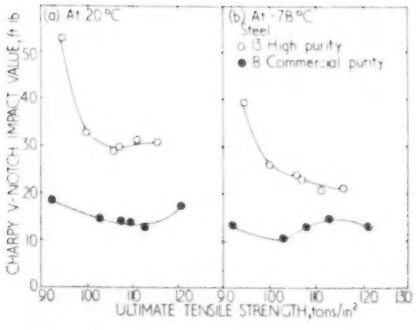
EXPERIMENTAL PROCEDURE

The high-purity steels (Table II, nos.1–3, 11–16) were prepared from carbonyl iron, nickel pellet, electrolytic chromium, molybdenum, and ferro-vanadium. Analyses of the raw materials are given in Table I. Carbon was added as graphite. These materials were melted by high-frequency induction, as 25-lb heats, *in vacuo* at a pressure of 1 micron, and were vacuum-cast to give 3-in. dia. ingots.

The commercial-purity steels were induction-melted in air, as 200-lb heats, with the exception of nos.4, 5, and 10, which were prepared as 25-lb heats. Analyses of the steels are given in Table II.

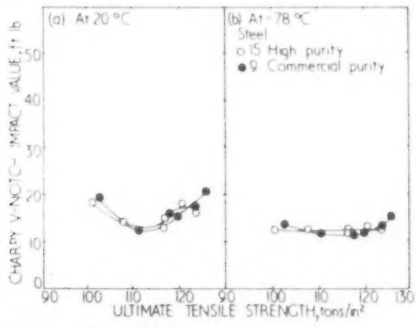
The steels, in the form of round bar, were oil-hardened, tempered for 1 h at temperatures in the range 200–500°C, and water-quenched. Tensile, impact, notched-tensile, plain-bar, and notched-fatigue tests were made at room temperature. Impact properties were determined also at −78°C.

The tensile properties were determined on 0·252-in. dia. standard test-pieces; Charpy V-notch test-pieces were used for the impact tests, and normal size or



a at 20°C *b* at −78°C

4 *Room-temperature and sub-zero impact properties of commercial-purity and high-purity* 4½%Ni–Cr–Mo–V *steels*



a at 20°C *b* at −78°C

6 *Room-temperature and sub-zero impact properties of commercial-purity and high-purity* 4½%Ni–Cr–Mo–V *steel containing 1·3–1·5%Si*

miniature Wöhler specimens for fatigue testing. Notch sensitivity was determined on plain and notched 0·357-in. dia. tensile test-pieces: the notch, made circumferentially in the middle of the gauge length, had a root radius of 0·005 in. and a 60° angle. The diameter of the test-piece at the root of the notch was 0·252 in.

TENSILE AND IMPACT PROPERTIES

Influence of carbon content in high-purity steels

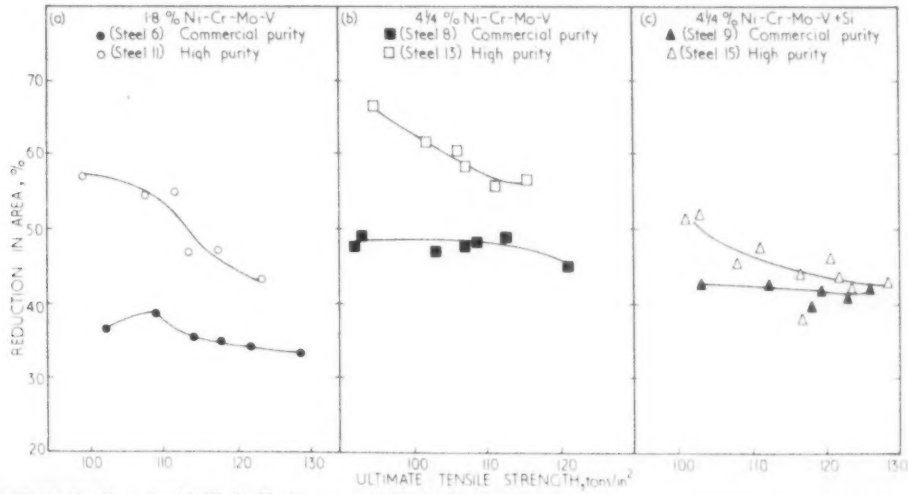
It is well known that carbon influences ductility, and that, for a given strength level, impact resistance decreases with increasing carbon. For commercial steels, however, the influence of carbon is, to a certain extent, masked by the occurrence of '350°C-embrittlement'.

In the relationships between tensile strength and room-temperature impact-resistance for 1½%Ni–Cr–Mo steels of high purity (Fig.2) the influence of carbon is revealed in a striking manner, since, owing to the absence of impurities, the steels are free from '350°C-embrittlement'. As the carbon is increased from 0·28% to 0·42% and to 0·54% there is a substantial overall depression of impact values, an effect which poses a problem in development of ultra-high-strength materials from conventional low-alloy steels.

The extent to which impurities normally present in commercial steels influence impact properties can be seen by comparing the curves for the high-purity steels 1 and 2 with those of steels 4 and 5, which were similar but were melted in air, from commercially-pure materials.

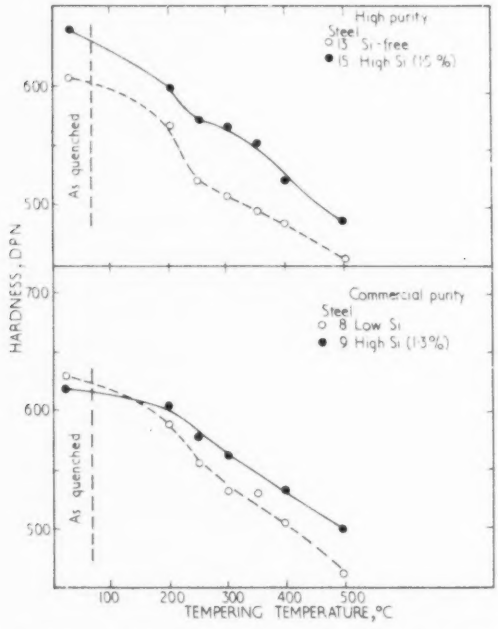
The fractures of the broken impact specimens showed important differences which are characteristic features of steels of the two grades of purity. Of the two steels containing 0·28%C, the high-purity type (no.1) gave completely fibrous fractures after tempering at temperatures above 350°C, whereas in the commercial-purity version, steel 4, fracture was not wholly fibrous unless the temperature was above 450°C. Further, the brittle fractures of the specimens tempered below these temperatures differed: the high-purity steel showed cleavage-type fracture, whereas its commercial-purity counterpart exhibited a considerable degree of intergranular fracture.

Fractures of lightly tempered, fine-grained steels are somewhat difficult to interpret, and estimation of the amounts of the respective types of fracture becomes increasingly debatable as the tempering temperature is lowered and, in general, as the carbon content is increased. The fractures of the 0·42%C steels


(a) 1·8 % Ni-Cr-Mo-V
● (Steel 6) Commercial purity
○ (Steel 11) High purity
(b) 4¼ % Ni-Cr-Mo-V
■ (Steel 8) Commercial purity
□ (Steel 13) High purity
(c) 4¼ % Ni-Cr-Mo-V +Si
▲ (Steel 9) Commercial purity
△ (Steel 15) High purity
REDUCTION IN AREA, %
ULTIMATE TENSILE STRENGTH, tons/in²

a 1·8%Ni–Cr–Mo–V *b* 4½%Ni–Cr–Mo–V *c* 4½%Ni–Cr–Mo–V + Si

5 *Influence of purity on tensile ductility of high-strength steels*



7 *Influence of silicon on tempering characteristics of* $4\frac{1}{4}$%Ni–Cr–Mo–V *steel*

(no.2, high-purity, no.5, commercial-purity) indicated a similar relationship with regard to ductile and brittle fractures, and a closer study of these is being made with the aid of the electron microscope.

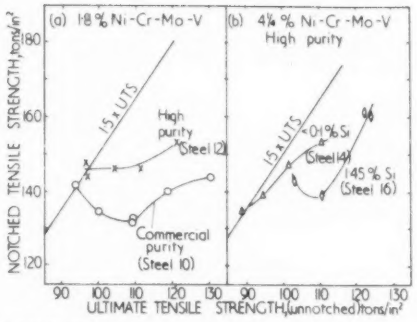
It can be concluded that at least a part of the improvement in impact properties gained from high purity, results from suppression of intergranular brittle fracture and the lowering of impact-transition temperatures.

Influence of purity on tensile–impact relationships in high-strength steels

It is in the field of ultra-high-strength steels that the advantages shown by steels of high purity appear to be of most practical interest; where impact strengths are inevitably very low, an improvement of even a few ft/lb becomes significant. It was therefore considered desirable to study the influences of purity on the properties of some types of established high-tensile steels.

The first type investigated was the $1\frac{3}{4}$%Ni–$1\frac{1}{4}$%Cr–1%Mo–$\frac{1}{4}$%V steel known as BACE 165, which was initially developed as a nitriding steel, and which, on account of its resistance to tempering, has more recently attracted attention as a high-strength material for aircraft applications. With a carbon content of 0·40–0·45%, this steel is capable of attaining a tensile strength of at least 100 tons/in² after tempering at 500°C.

By removal of impurities from this steel significantly increased values of room-temperature toughness were obtained, particularly when the high-purity steel was tempered to strength levels below about 120 tons/in² (Fig.3*a*), whereas the impact values of the comparable commercially pure material (steel 6),



a 1·8%Ni–Cr–Mo–V *b* $4\frac{1}{4}$%Ni–Cr–Mo–V high-purity

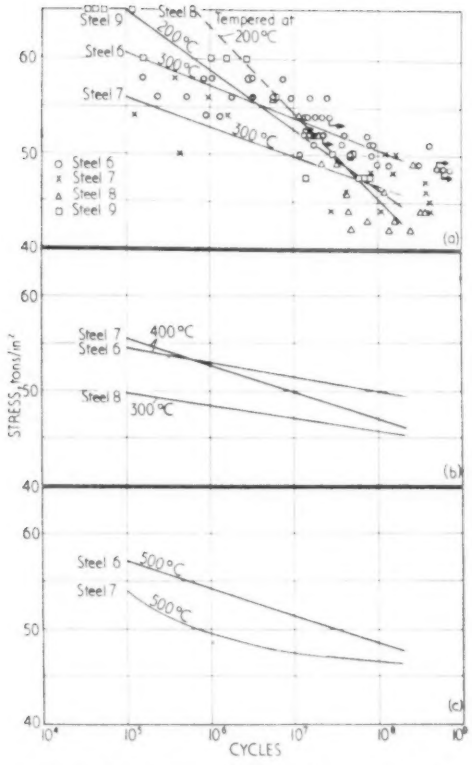
8 *Notch sensitivity of high-strength steels*

remained virtually independent of tempering temperature between 200° and 500°C. The advantage of high purity extended to properties at sub-zero temperatures, but at −78°C the amount of improvement at the lower strength levels was somewhat reduced (Fig.3*b*).

Results for a commercially pure steel similar to steel 6 but having a lower molybdenum content (0·49%) are also included in Fig.3. The lower-molybdenum steel had slightly improved impact properties at room temperature, but at the higher strengths this improvement was not maintained at −78°C. It would appear that in the more lightly tempered material, increase in molybdenum content from $\frac{1}{2}$% to 1% changed the slopes of the impact/transition curves, and may also have lowered the transition temperatures. It is presumably as a result of its high molybdenum content that the impact values of the commercially pure BACE 165 of standard composition did not pass through a minimum at an intermediate tempering temperature.

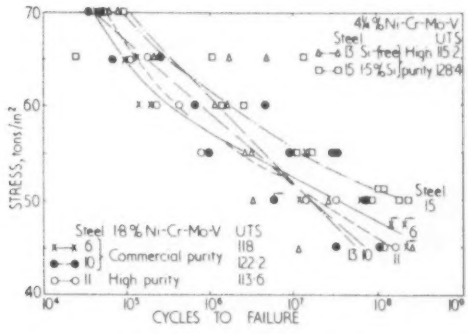
The second type of high-strength steel investigated was the $4\frac{1}{4}$%Ni–Cr–Mo–V steel (vanadium-modified BS En 30B). In this instance, the elimination of impurities resulted in a remarkable improvement in toughness, both at room temperature and at −78°C (Fig.4). As the carbon contents of these steels (nos.8 and 13) were somewhat lower than those of the previous steels (nos.6, 7, and 11), the magnitude of the improvement confirms the impression that the influence of impurity elements is dependent on the carbon content.

As regards tensile properties, the data which have been accumulated suggest that the purity of a steel has a slight influence on its ultimate tensile strength. For a given carbon content and a given tempering temperature, the strength of high-purity steel may be lower than that of the corresponding commercial-purity steel, by up to about 4 tons/in²: for several pairs of steels examined the difference remained constant at about 4 tons/in² for a range of tempering temperatures. There is thus no indication that a high-purity steel tempers more rapidly than a commercial steel. The lower strength of high-purity steels is not entirely surprising in view of the fact that commercial steels generally contain 0·2–0·3%Si and about 0·5%Mn, as residual deoxidizers, and that these elements have fairly strong solid-solution hardening effects in ferrite,[16] which could account for the observed difference in tensile strength.



Heat treated to about *a* 120 tons/in² *b* 110 tons/in² *c* 100 tons/in²

9 *Fatigue properties of commercial-purity high-tensile steels*



10 *Comparison of plain-bar fatigue properties of high-purity steels and commercial-purity steels, tensile strength about 120 tons/in²*

Tensile ductility, as measured by percentage elongation, was about 1–2% higher in high-purity steels. Reduction of area, on the other hand, is a more sensitive index of ductility, and the better values observed with high-purity steels (Figs.5*a* and *b*) are perhaps more striking than the influence of purity on impact properties, already discussed. Figure 5 shows that the strength dependence of ductility shown by the high-purity steels is not shared by the commercial-purity steels, and the difference between the reduction-of-area values of the high-purity and commercial-purity steels decreased as the tensile strength increased. For the high-purity 1·8%Ni–Cr–Mo–V steel, reduction of area still amounted to about 44% at a strength level of 120 tons/in² and in the 4¼%Ni–Cr–Mo–V steel it was 56% at 115 tons/in².

Influence of silicon

Reference has already been made to the significance of silicon as a major alloying element in ultra-high-tensile steels, as a means of retarding softening during tempering.

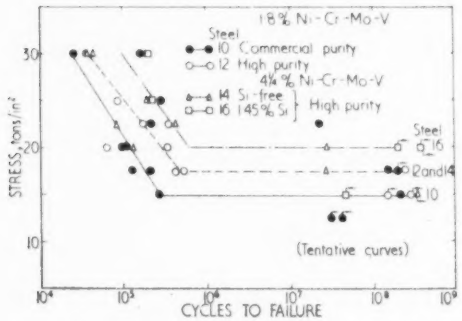
In the present investigation, the influence of silicon was studied in steels of the 4¼%Ni–Cr–Mo–V type, using both commerical- and high-purity materials (nos.9 and 15). The results (Fig.6) show that in the presence of 1·3–1·5%Si there was virtually no difference in impact properties between steels made with pure materials and those of commercial purity. Silicon retarded tempering to a certain extent (Fig.7), and its embrittling influence, produced a pronounced trough in the curve of room-temperature impact values for the high-purity steel (Fig.6): minimum values occurred after tempering in the region of 400°C. Thus, although the presence of silicon tends to raise the tempering range in which embrittlement occurs, and allows steels to be tempered at higher temperatures without undue loss of strength, the use of this element may not be advantageous if resistance to tempering and avoidance of embrittlement can be obtained by other means.

A further disadvantage of silicon is its deleterious influence on tensile ductility, as measured by reduction of area (Fig.5*c*). This effect was only slight in the case of the commercial-purity 4¼%Ni–Cr–Mo–V steel, but for the high-purity steel the addition of 1·5%Si reduced the general level of reduction of area to values below those of the commercially pure low-silicon steel. Although there is a certain amount of scatter in the results, the values for the high-purity, high-silicon steel appeared to be little better than those for the comparable commercial-purity steel.

NOTCH SENSITIVITY OF HIGH-STRENGTH STEELS

As the tensile strength of quenched-and-tempered steels is increased above 100 tons/in² the usefulness of the notched-impact test, in distinguishing between different steel compositions and treatments, becomes considerably less. This is due to the inherently low impact values associated with lightly tempered martensite. Notch sensitivity, as determined by the notched tensile test, usually manifests itself only at high strength levels. (Steels are said to be notch sensitive if the ratio of notched-tensile strength to (un-notched) ultimate tensile strength is less than about 1·5.)

Results of notched-tensile tests on selected commercially pure and high-purity steels, after tempering at suitable temperatures, are summarized in Fig.8. According to Neuber's formula, the theoretical stress-concentration factor in the notched test-piece employed was 4·5. At strength levels above 100 tons/in² all the steels examined showed notch sensitivity in varying degree. For the high-purity 1·8%Ni–Cr–Mo–V steel (no.12) the notched strength increased with rise in un-notched tensile strength (Fig.8*a*) whereas in the corresponding commercially pure steel (no.10) notched strength was lower and showed a distinct minimum



11 *Comparison of notched-fatigue properties of high-purity and commercial-purity steels, tensile strength about 120 tons/in²*

after tempering in the region of 400° C. This behaviour is interesting in view of the absence of a 'trough' in the curve of impact value/tensile strength for this type of steel (Fig.3): it appears to imply that notched strength is more sensitive to impurity elements than is impact resistance. A similar 'trough' in notched-strength values resulted from the addition of 1·45%Si to a high-purity 4¼%Ni–Cr–Mo–V steel (Fig.8*b*). From the point of view of applications at high strength levels, this effect constitutes a further disadvantage resulting from the use of considerable amounts of silicon in high-purity steels.

FATIGUE PROPERTIES OF HIGH-STRENGTH STEELS

Room-temperature fatigue tests were carried out on rotating-bend machines of NPL design, running at 5000 rev/min and using Wöhler-type test-pieces such that the bending stresses were uniform over their gauge lengths. Two sizes of test-piece were used, according to the amount of material available: for the commercially pure steels (nos.6–9) the test-piece was of normal size, having a parallel 1-in. gauge length of 0·275 in. dia.: for the remaining steels (nos.10–16) a miniature test-piece was employed, of 2-in. overall length, with a 0·178 in. dia. gauge length and a parallel portion 0·250 in. long. The miniature samples were inserted in closely fitting shackles, to adapt them to the machines used for the larger test-pieces.

Preparation of the fatigue specimens involved heat-treatments in the rough-machined condition (0·030 in. being allowed on the diameter for decarburization), after which they were turned to the final dimensions, and polished by normal procedures. The steels examined by plain-bar fatigue tests were the same as those used for determination of tensile and impact properties. The commercial-purity steels nos.6–9 were tested at strength levels of 120, 110, and 100 tons/in². The results (Fig.9) indicated that, on the basis of such tests (i.e. for stresses resulting in fatigue lives of up to 2×10^8 cycles), at high strength levels, the steels did not possess fatigue limits. In view of the scatter of results, the fatigue properties were, in most cases, best represented by straight lines.

The strength levels at which the steels were tested had a relatively small influence on the results. As tempering temperatures were raised, i.e. as tensile strengths were lowered, *finite fatigue* lives at high stresses were somewhat reduced, but at lower stresses this effect was negligible, so that for lives of 2×10^8 cycles the initial tensile strength of the materials (within the range 100–120 tons/in²) did not appear to affect fatigue strength. The tests revealed no consistent influence of steel composition.

It will be recalled that the commercially pure steels nos.6–9 were cast as 200-lb ingots, and the high-purity steels as 25-lb ingots. Since in processing the commercial- and high-purity steels to bars of $\frac{7}{8}$ in. and $\frac{5}{8}$ in. dia. respectively, the two groups of steels received different degrees of working, it was necessary to provide some means of taking this difference into account when comparing the two sets of fatigue data, and also to assess the influence of test-piece size. For this purpose, steel 10 was made as an air-melted commercial-purity heat, and cast in the same size of ingot mould as was used for the vacuum-melted high-purity steels.

To establish the correlation between the fatigue properties determined on normal and on miniature test-pieces, a further quantity of steel 6 was heat-treated to 120 tons/in² and tested in the form of miniature specimens. The results confirmed that size of test-piece had no influence on fatigue properties, and the values obtained for this steel, and for steel 10, were subsequently used to compare the fatigue properties of the commercial-purity and high-purity steels. To study the influence of purity, the commercially pure steel no.10 and the high-purity steels nos.11, 13, and 15, were heat-treated to a nominal tensile strength of 120 tons/in². The fatigue results, *see* Fig.10, showed considerable scatter at some stress levels, and the unavoidably small scale of the melts of individual steels limited the number of tests which could be made at each stress level: it is, nevertheless, clear that removal of impurities had no significant influence on fatigue properties. In view of the claims made for the improved fatigue resistance of vacuum-melted steels, this result is rather interesting, suggesting (as had been suspected) that the advantages previously claimed for vacuum-melted steels were primarily the result of lower inclusion content, a feature unlikely to be so sharply defined when relatively clean laboratory-made steels were compared.

Each of the high-purity steels was heat-treated to a nominal 120 tons/in² strength level, as gauged by pilot hardness tests, but the actual figures ranged from 113·6 to 128·4 tons/in². Since, as already observed, it was only at the *higher* stresses that the strength level had a significant influence on fatigue properties, these variations in strength level could be ignored when making comparison with commercial steels. The only significant effect of composition which emerges from Fig.10 is that the high-silicon 4¼%Ni–Cr–Mo–V steel (no.15) had better properties than the remaining steels.

In a further attempt to elucidate the influence of impurities on fatigue properties, steel 10 and steels 12, 14, and 16 (duplicate melts of nos.11, 13, and 15) were subjected to notched-fatigue tests, again at a nominal tensile-strength level of 120 tons/in². The test-piece used was similar to the miniature specimen employed in the plain-bar fatigue testing, except that the parallel portion was left oversize, and was provided with a central circumferential 60° notch of such a form that the root diameter was 0·178 in. Root radius of the notch was 0·005 in., giving a stress-concentration factor of 3·0. The results (Fig.11) showed some scatter, but the notched-fatigue tests appeared to

show that in this case purity exerted some influence: the properties of the high-purity steel no.12 were superior to those of its commercial-purity counterpart (no.10). The high-silicon 'pure' $4\frac{1}{4}$%Ni–Cr–Mo–V steel exhibited slightly better fatigue properties than the silicon-free steel, again indicating a possible advantage of high silicon.

In addition to bringing out the influence of composition, notching of the specimens greatly reduced fatigue strength and introduced an endurance limit for each of the steels tested.

DISCUSSION

A primary object of the present investigation was to show the influence of impurities on the mechanical properties of some lightly tempered martensitic steels. Several effects were observed, and it has been demonstrated that high-purity steels offer significant improvements in impact properties, in tensile ductility, and in notch sensitivity as indicated by notched tensile tests. The study did not include determinations of the effects of individual impurities, but, from evidence previously obtained,[2] it seems probable that the most deleterious trace elements (when present at the levels at which they commonly occur in commercial steels) are phosphorus and nitrogen. The presence of antimony, arsenic, or tin would also be expected to have a significant effect.

Earlier work has also shown the deoxidizing elements silicon and manganese to have an embrittling influence, but their effects are small when they are present at normal levels. The investigation now reported has confirmed that larger amounts of silicon have appreciable effects on the impact properties of high-purity steels and has indicated the harmful influence of that element on the tensile ductility and notch sensitivity of these materials. In the commercially pure steels examined, impact- and reduction-of-area values for the high-silicon steels were not significantly different from those for steels of low silicon content, a finding which is in agreement with the work of Shih *et al.*[7] on steels of the SAE 4340 type.

Electron-micrographic studies threw some light on the influence of silicon on tempering, revealing that the presence of 1·5%Si considerably suppressed carbide precipitation during tempering. Although some difficulty was experienced in obtaining satisfactory carbon-extraction replicas from the high-purity high-silicon steel (no.15), it appeared that there was very little precipitation after tempering at 300 C, and even after tempering at 500 C there was much less precipitation than in the silicon-free $4\frac{1}{4}$%Ni–Cr–Mo–V steel (no.13); the individual particles were largely spheroidal, in contrast to the platelet carbides observed in low-silicon steels.

It is unfortunate that silicon at the level of 1·5%, which has been found useful in retarding the softening of martensite on tempering, impairs the mechanical properties of high-purity steels to an extent which appears to be equivalent to the *combined* effects of the trace impurities normally present in commercial steels. Thus, while high-purity steels may be of interest for certain high-strength applications, the desirability of utilizing silicon as a means of modifying their tempering characteristics appears to be ruled out. However, the freedom of high-purity steels from susceptibility to '350°C-embrittlement' renders it unnecessary to resort to higher silicon for steels intended for service requirements necessitating the use of tempering temperatures of 300–325 C without impairment of impact properties. Obviously, it is still necessary to ensure that the high-purity steels provide adequate strength after tempering at such temperatures, and in this connexion the role of carbon, as revealed in the present investigation, is of great importance. While, however, higher strengths may be obtained by increasing the carbon content, the detrimental effects of carbon on impact properties is a restrictive factor.

The undesirability of employing high silicon or carbon contents in high-purity steels, as a means of securing increased strength after tempering at relatively high temperatures, indicates the desirability of achieving this end by other means. In the present work partial success was attained by using high molybdenum contents and/or vanadium. Electron-microscopical examination of the steels confirmed that the high-molybdenum 1·8%Ni–Cr–Mo–V steel showed increased resistance to tempering: after treatment at 500 C this steel appeared to be comparable, with regard to carbide size and distribution, with the lower-molybdenum $4\frac{1}{4}$%Ni–Cr–Mo–V steel tempered at 300 C.

Although high-purity steels at high strength levels offer certain distinct advantages compared with commercial-purity materials, they would be still more attractive if they provided useful improvement in fatigue properties. Conflicting claims have been made regarding the effects of vacuum-melting on fatigue properties: in some instances improvements have been obtained,[12,13] while in others little difference has been found between the properties of air- and of vacuum-melted steels.[17] It appears probable that the influence of melting procedure on inclusion content can exert important effects on fatigue properties but that such differences may not be revealed in comparison of air- and vacuum-melts from laboratory induction furnaces. The results obtained showed that elimination of impurities had no significant effect on un-notched fatigue properties, but notched fatigue test data suggested a small beneficial effect of purity. This finding, when considered in relation to the improvements in notch toughness and notched-tensile strength shown by high-purity steels, indicates the desirability of more detailed study of the notch-fatigue properties of high-purity steels.

CONCLUSIONS

1. In high-strength Ni–Cr–Mo–V steels the removal of normal impurities results in improved impact properties, tensile ductility, and notched-tensile strength.

2. Increasing the carbon content of high-purity steels has a markedly deleterious influence on impact properties: this effect is more clearly revealed by the high-purity materials than by steels containing impurities at normal commercial levels.

3. The addition of about 1·5%Si modifies the tempering characteristics of martensite, giving some practical advantage in steels of commercial purity, but in high-purity steels this amount of silicon has detrimental effects on impact properties and tensile ductility, and such steels show no advantages over commercial-purity types.

4. Plain-bar fatigue tests on high-strength steels revealed no endurance limits, and the results indicate that the finite-life fatigue strengths are not affected by the purity of steels heat-treated to about 120 tons/in^2. In contrast, in notched-bar fatigue tests, steels at this strength level show an endurance limit. The preliminary data already established indicate that high-purity steels are slightly superior to those of commercial purity as regards both finite-life fatigue stress and endurance limit.

ACKNOWLEDGMENT

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Correspondence on the paper

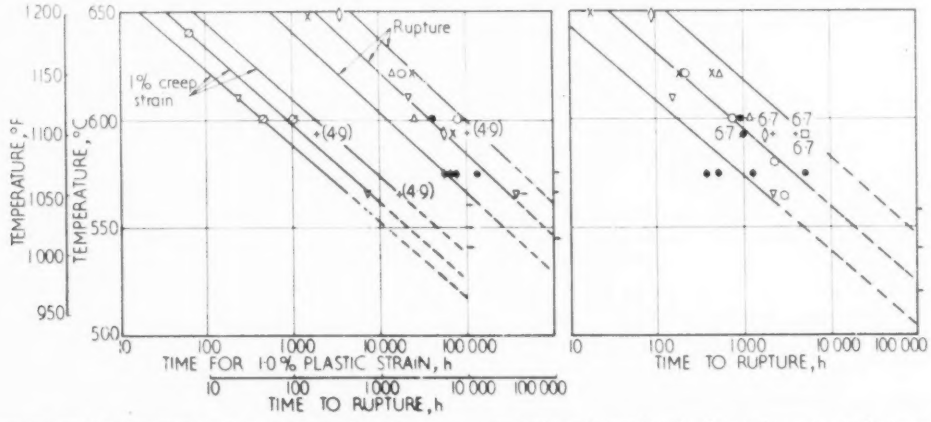
The creep and rupture properties of 2¼% chromium–1%molybdenum quality steel*

Mr **L. M. T. Hopkin** and Mr **M. F. Day,** of the National Physical Laboratory, Teddington, Middlesex, wrote: The compilation of British data on the creep and rupture properties of the 2¼%Cr–1%Mo steel by Murray *et al.* implies that design stresses at the service temperatures of 538°C (1000°F) and 565°C (1050°F) can be considerably above those recommended in the ASME Boiler Code. Although the British data may be more extensive than those on which the ASME Boiler Code is based, the compilation of British data should be critically examined before any recommendation for working stresses higher than those in general use at present can be justified.

At both the temperatures in question the bulk of the results are for times of less than about 5000 h; only four rupture test results at 538°C and two rupture test results at 565°C are for greater times up to about 20000 h. The results for 1% creep strain are no more numerous. These results have been extrapolated by the authors on a curved relationship between log stress and log time to 1% creep strain and to rupture to estimate the stresses required for these criteria at 100000 h. The doubt which exists with any extrapolation is obviously aggravated when such long extrapolations are made on curved relationships.

As a general principle, we believe that the results of extrapolations of both log stress/log time, and temperature/log time

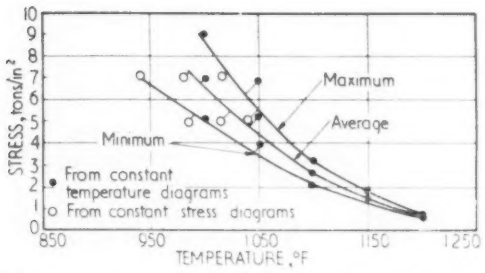
* J. D. Murray *et al.*: *JISI*, 1959, **193,** Dec., 354–359.



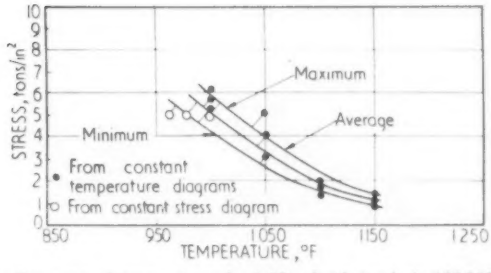
Figures in brackets indicate tests at a stress other than (i) 5 tons/in^2 (ii) 7 tons/in^2

← indicate interpolated values (min., max., and average) from paper by Murray *et al.*

A 2¼%Cr–Mo *quality steel: temperature* v. *time to rupture at constant stress,* (i) *5 tons/in*^{2} (ii) *7 tons/in*^{2}



B *Rupture properties at 100000 h, 2¼%Cr–Mo quality steel*



C *Stresses and temperatures for 1·0% plastic strain in 100000 h, 2¼%Cr–Mo quality steel*

plots should be considered together to obtain the stress/temperature relationships at 100000 h such as are shown in Figs.10 and 11 of the paper. Both methods of extrapolation are open to some criticism but each method has its individual merits. The results of extrapolations by one method are, therefore, a check on the results from the other. This technique was employed successfully in the compilation of data for another steel.*

Contrary to Murray *et al.*, we consider that use could have been made of the temperature/log time extrapolations in the present paper. It is agreed that fewer results are available for the temperature/log time than for the log stress/log time plots but this fact is offset by the greater reliability of the temperature/log time extrapolations. This is so since the temperature/log time relationship is more nearly linear than that for the log stress/log time plot.

Examination of the log stress/log time plots shows that there are results for stresses of 4·9 and 6·7 tons/in² which could be included in the temperature/log time plots for 5 and 7 tons/in², respectively (Figs.5 and 6). The log stress/log time plots indicate that the errors in time of these additional points, compared with results at the nominal stresses, are about 1/20th of a logarithmic cycle and can be neglected. Figures 5 and 6 have been replotted here in Figs.*A* (i) and (ii) together with the additional results. Also in Fig.*A*(i) are plotted results for 1% strain at 5 tons/in²; the number of points being the same as that used by the authors to establish the 1% strain curve at 593 C. It can be seen in both Figs.*A*(i) and (ii) that the maximum, minimum, and average values of the 100000-h properties interpolated from the curves in Figs.10 and 11 are consistently above any reasonable set of lines through the test results. With regard to the slope of the lines shown in Figs.*A*(i) and (ii), it should be pointed out that with increasing temperatures of test, particularly over 600 C, thinning of the cross-section of the test pieces as the result of heavy oxidation may have made the times for 1% strain and rupture shorter than they would otherwise have been. Allowance for this effect, therefore, would change the slope of the lines so that even lower temperatures would be obtained on extrapolation to 100000 h. When the extrapolated values from Figs.*A*(i) and (ii) are added to Figs.10 and 11 in the paper as shown in Figs.*B* and *C*, these points are below those obtained from the log stress/log time extrapolations. This discrepancy must be accepted as being due to insufficient results of long-time tests.

Interpolated values from smooth curves drawn through all the points for both methods of extrapolation in Figs.*B* and *C* have been used to compile Tables 1 and 2. (cf. Tables III and IV of the paper). It will be seen that the 100000-h stress values at 1000 F and 1050 F are somewhat lower than suggested in Table III by the authors.† Also, the stresses derived from the British data at 1000 F are now only slightly above those recommended in the ASME Boiler Code while at 1050 F there is no significant difference. It appears, therefore, that there is no justification from these results for using higher working stresses than at present.

We would like to point out an editorial error in that the symbols for the numerous samples tested which are shown in the key to Table I do not correspond with the symbols used in the graphs. Also, Figs.10 and 11 contain two and three points, respectively, at about 1110 F which cannot have been obtained from any diagrams appearing in this paper.

* W. H. Bailey *et al.*: *Proc. IME*, 1957, **171**, 911.

AUTHOR'S REPLY

Mr **J. D. Murray** (The United Steel Companies Ltd) wrote in reply: The data given in the paper published in December 1959 presented the position of British work at the end of 1957. Since that time tests of longer duration have been completed and these have shown that the values extrapolated for temperatures of 1000 F and 1050 F were optimistic. Obviously, therefore, the comments made by Hopkins and Day are of importance. At the same time it must be pointed out that their thesis is based on an incorrect premise which is that the temperature/log time relationship is approximately linear and is more reliable than the log stress/log time plot. Glen has clearly shown that the temperature/log time relationship is not linear. Further, the reliability of this method has never been proved by experimental data. The main attraction of the method is that it takes into account thermal action, i.e. it assumes that structural deterioration is a simple time/temperature relationship. Much of the recent metallurgical work done on the tempering behaviour of alloy steels has shown that the size and distribution of alloy carbides are two of the main factors governing tempering resistance. Such work has also shown that the size and distribution of these carbides are greatly affected by the temperature at which they form and, in general, the lower the temperature of formation the finer and more numerous are the particles. It has also been shown that creep properties are affected by the size and distribution of carbide particles and once again fine and numerous particles are of advantage. Consequently, it would be expected that carbides formed at lower temperatures will be better from the point of view of creep, than carbides formed at higher temperatures. If we consider those carbides formed during testing and which are known to have a large effect on the creep properties we see that if we accelerate our test durations by testing at higher temperatures, then we are likely to form a type of carbide that will give a pessimistic result. In addition, if the temperature of testing is so high that rapid oxidation is occurring a still more pessimistic result will be obtained.

When preparing the paper that was published in December 1959 much of the above reasoning was considered and the temperature/log time relationships were rejected because of the difficulty in constructing correct curves. Hopkins and Day, with a little more information at their disposal, have now attempted this. Nevertheless, if their Figs.*A*(i) and (ii) are looked at closely the speculative nature of their attempt will

† Since this letter was drafted, Murray presented to the International Discussion on the Long-Term Behaviour of High-Temperature Steels, Dusseldorf 1960, a revised form of his paper, incorporating additional results of long-time tests, in which the stresses for rupture in 100000 h are substantially in agreement with those presented in Tables 1 and 2. This is vindication of the method used in the present letter of evaluating long-time creep and rupture properties.

TABLE 1 100000-h stress data

Temp., °F	Stress, tons/in² 1% plastic strain Max.	Min.	Average	Rupture Max.	Min.	Average
1000	6·0	4·3	5·1	8·7	4·9	6·5
1050	4·0	2·7	3·3	5·5	3·4	4·4
1100	2·4	1·6	2·0	3·2	2·1	2·6
1150	1·4	0·9	1·2	1·8	1·3	1·5
1200	...	...	...	0·8	0·6	0·7

be realized. Taking Fig. 4(i), within the 18 rupture test results that are shown there are 8 variables which comprise different heat treatments and different cast numbers. There is no one set of results from one cast with one heat treatment which gives a definite trend of the slope of the temperature/log time plot. In constructing their mean curve they seem to have placed more reliance on four results which give lower values than on a greater number of results which have given higher values, i.e. they have more points above the average line than below it. In addition, the majority of the lower values they have included were obtained from a heat treatment which

TABLE 2 Comparison of stresses obtained with the ASME Boiler Code values

Temp., °F	ASME boiler code, lb/in²	Stresses derived from British data, lb/in² 60% average stress for rupture in 10^5 h	80% min. stress for rupture in 10^5 h	Stress for 1% plastic strain in 10^5 h
1000	7800	8750	8800	9600
1050	5800	5900	6100	6000
1100	4200	3500	3800	3600
1150	3000	2000	2300	2000
1200	2000	940	1100	...

is not considered a practical one. It can well be argued, therefore, that their average lines should be a little higher and, in fact, they could be of a less steep slope. This would give higher values than they have shown.

Possibly the main point that Hopkins and Day have brought out is the divergence of opinion in the UK on the presentation and extrapolation of creep data. Consequently, any contributions that are made to this subject are of great value and because of this, their contribution is acknowledged with thanks.

Dislocations in α-iron

D. G. Brandon, M.A., Ph.D., and J. Nutting, M.A., Ph.D., B.Sc.

SYNOPSIS

Electron optical examination of the dislocation distribution in pure α-iron after different amounts of plastic deformation shows that irregular sub-boundary networks are formed. In coarse-grained material these sub-boundaries are partially responsible for work hardening.

1826

INTRODUCTION

ON EXAMINING METAL FOILS by transmission in the electron microscope it is possible to observe directly the dislocation distribution. Experiments have now been carried out to determine how the dislocation distribution in α-iron changes as a result of plastic deformation. An attempt has then been made to correlate these observations with the work-hardening characteristics of the iron.

It has been found that with increasing plastic strain the dislocations form sub-boundary arrays as previously predicted from microbeam X-ray diffraction by Gay and Kelly.[1] Similar electron optical observations have been made on aluminium and stainless steel by Hirsch and co-workers.[2, 3] Kelly[4] and Ball[5] have suggested that the work-hardening characteristics of α-iron are related to the sub-grain size. But Hyam and Nutting,[6] and Irvine and Pickering,[7] have shown that the ultimate tensile stress of steel is greatly dependent upon the initial ferrite grain size. Thus Kelly's and Ball's views can only be reconciled with this fact if it is assumed that sub-grain size is related to the grain size. The present experiments have shown that this is not the case and that sub-grain diameters of less than 1–2 μ are not observed. It is now suggested, therefore, that work hardening in fine-grained material is chiefly due to forest interaction of dislocations and is not very severe, while in coarse-grained material sub-grain formation can occur readily and this leads to an extra hardening increment, although the yield stress is low.

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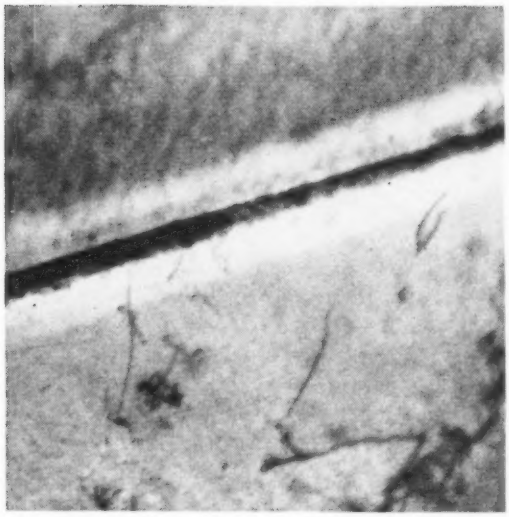
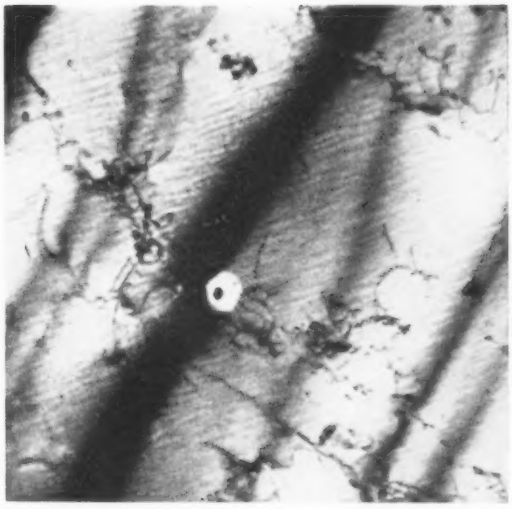
EXPERIMENTAL

Iron having a purity of 99·97% (BISRA reference no.AHP19) was used for this investigation, the analysis follows: 0·0024%C, <0·001%Si, 0·004%Mn, 0·005%S, <0·001%P, 0·01%Ni, 0·001%Cr, 0·004–0·007%Cu, <0·001%Al, 0·0022%O, 0·0013%N 0·0000005%H. No traces of Mo, W, V, Ti, Co, or Sn were found. The bar was cold-rolled from $\frac{3}{8}$ in. to give a foil 120 μ thick. Although this may be less than the ferrite grain size of annealed material, it was found that the stress/strain curve corresponded to that of the metal in a more massive form.

To study the very early stages of deformation specimens of decarburized foil were prepared by annealing for 48 h at 850 C in moist hydrogen and after this treatment no yield point could be detected. To investigate the relationship between the surface slip markings and the dislocation distribution a foil containing a large single crystal was prepared using the strain anneal process of Edwards and Pfeil.[8]



1 *Grain boundary showing patchy contrast and 'clear' zones* ×30000



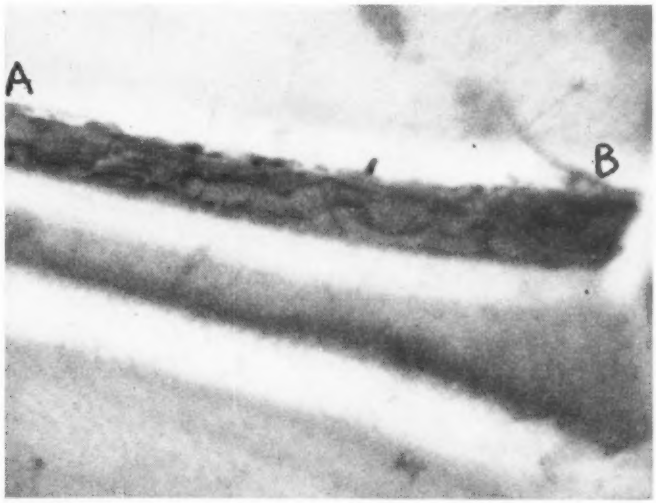
3 *Dislocation networks in coarse-grained foil deformed to a stress of 6 tons/in² (13% strain). The striations running horizontally across are a surface effect produced by electropolishing* ×15000

The foils were first annealed *in vacuo* and after deforming thin films were prepared by electropolishing using the technique of Brandon and Nutting.[9] The foils were examined by transmission in the Siemens Elmiskop 1 electron microscope operated at 80 kV.

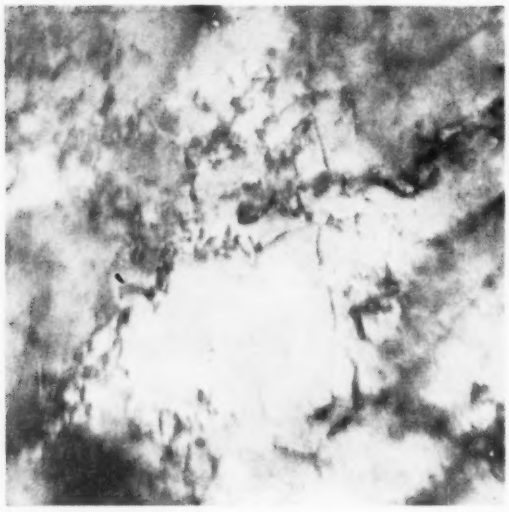
RESULTS

Metallographic features of annealed polycrystalline foils

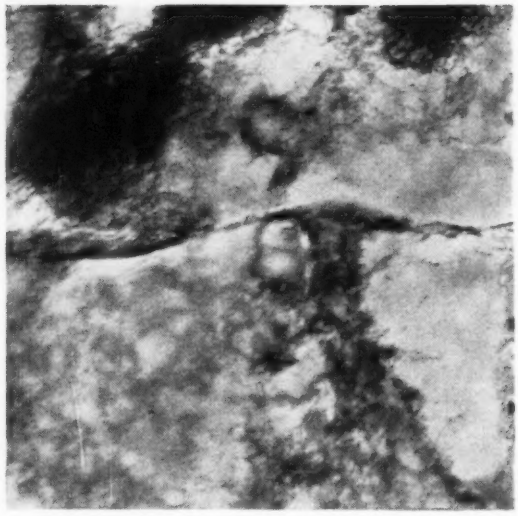
Very few dislocations were visible in annealed specimens, and it was estimated that within the grains the dislocation density was 10^6–10^7 cm^{-2}. Grain boundaries usually appeared as a fine double line where they intersected the top and bottom of the film. On both sides of the grain boundaries were regions which appeared more transparent to electrons than the rest of the grains (Fig.1). This effect does not seem to be due to preferential electropolishing in the grain boundary region (though this is sometimes found) but is thought to correspond to an impurity-free zone. The presence of impurities would be expected to increase the non-coherent scattering of the electron beam and hence to decrease the intensity of the transmitted beam. The grain boundary acts as a sink for impurities and this might lead to a transparent impurity-free zone adjacent to the boundary. Very often small loops of dislocation were observed beginning and ending at grain boundaries, and in or immediately adjacent to the boundaries it was possible to detect a network of fine lines (Fig.2). Dislocations have frequently been seen to emerge from grain boundaries and it is probable that the networks of fine lines are dislocations which are loosely bound to one side of the boundary.



2 *Grain boundary and adjacent dislocation network. A grain boundary runs from A to B and superimposed upon this are various dislocations forming an incomplete network. The other bands running parallel to the boundary are extinction contours* ×40000



4 *Sub-boundary formation in foil deformed to a stress of 2 tons/in² (8% strain)* ×40000



5 *Grain boundary kinks in foil deformed to fracture* ×40000

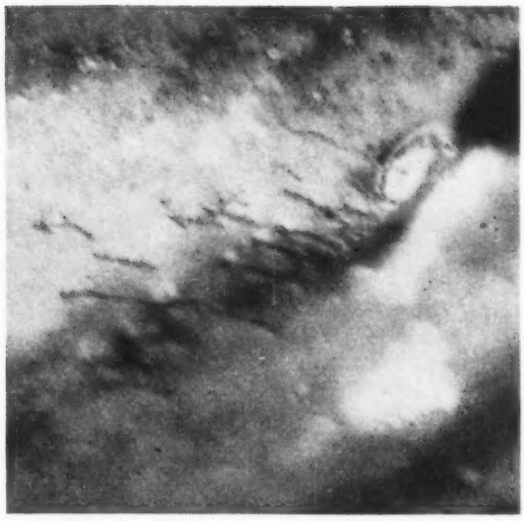
The grain boundary may thus be acting as a reservoir for dislocations, and under the influence of an applied stress, these dislocations may break free and contribute appreciably to the initial strain. The dark patches at the grain boundary (Fig.1) may be associated with regions of good and bad fit as suggested by Mott,[10] but no similar effects have been reported in fcc metals and it is more likely that the contrast is due to small precipitates or segregation of impurities. No evidence has been obtained for segregation or precipitation within the grains, although surface replicas strongly suggest that such an effect is present.[11]

Deformed polycrystalline foils

The smallest deformation which can be investigated in annealed iron is the yield point extension. This varied with the grain size, but was about 2 or 3%.

After yielding, the dislocations were arranged in loose networks which were frequently associated with sub-boundaries and inclusions (Fig.3). With increasing deformation the networks become more clearly delineated and the dislocations within the networks become more tightly packed (Fig.4). The dislocation density for a similar strain decreased with increasing grain size although in any one specimen the dislocation distribution was very heterogeneous. High dislocation densities were frequently noted on one side of a sub-boundary or grain boundary, and grain boundary kinking was sometimes found after heavy deformation (Fig.5). Similar effects on replicas were found by Brandon and Nutting.[11] The size of the sub-grains was about 2 μ, which agreed well with the cell size determined by microbeam measurements.[4] Where the plane of the thin film was almost parallel to the plane of a network it was possible to see the dislocation

6 *Dislocation network showing three-fold nodes at junction of* (111) *and* (100) *dislocations* ×40000



7 *Irregular pile-up at an inclusion in a decarburized foil deformed to a stress of 3·5 tons/in² (1% strain)* ×75000

8 *The straight dislocations form a sub-boundary. The irregular bowed-out dislocations are breaking through the sub-boundary. Single crystal foil deformed to a stress of 5·25 tons/in² (4% strain)* ×75000

arrangements, and many threefold nodes were observed (Fig.6). Since a threefold node is geometrically impossible if only ⟨111⟩ dislocations are present, this observation provides strong evidence for the production of ⟨100⟩ dislocations by a reaction such as; $\frac{1}{2}a[111]+\frac{1}{2}a[1\bar{1}\bar{1}]=a[100]$. If this reaction occurs without climb taking place the ⟨100⟩ dislocation will lie at the intersection of two glide planes. The direction along which the ⟨100⟩ dislocations in the networks lie thus provides information concerning the choice of slip plane. In Fig.6 there is no obvious set of parallel dislocations and it may be inferred that the dislocations which are interacting do not lie on any specific crystallographic plane.

By measuring the splitting of Kikuchi lines and the circumferential displacement of spots on diffraction patterns from regions containing networks it has been possible to obtain values for the component lattice misorientations either in the plane of the foil (Kikuchi lines) or perpendicular to the plane of the foil (diffraction spots). The misorientation across dislocation networks measured in this way varied greatly from grain to grain in any one specimen. However, estimates have been made of both the sub-grain misorientation and the dislocation density in several specimens and the results are presented in Table I.

Decarburized foils

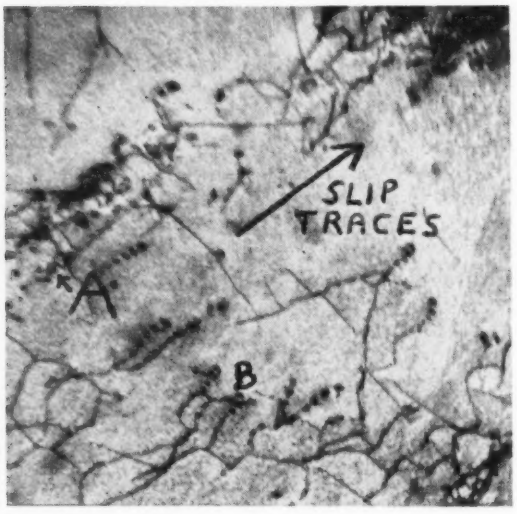
In order to investigate the early stages of network formation, dislocation arrangements were studied in decarburized foils which showed no yield extension and therefore could be deformed less than 3%. The flow stress increased rapidly to about 3·5 tons/in² after which the rate of work hardening decreased and became linear with strains greater than 2%. These changes are reflected in the dislocation arrangements observed inside the foils. After 0·8%, strain dislocations were found in irregular pile-ups at inclusions (Fig.7) and grain boundaries, while in some areas dislocation interaction was observed. After 2%, elongation networks, similar to those found in undecarburized specimens, were observed.

Deformation of a single crystal foil

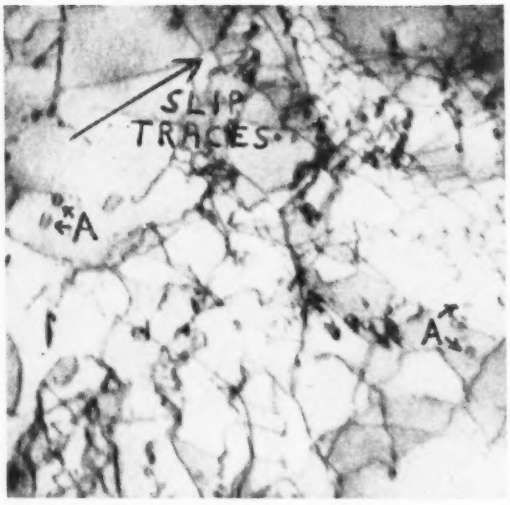
To determine the relationship between the slip bands found on the surface of a deformed specimen and the dislocation arrangements within the specimen, a single crystal foil was prepared by the strain-anneal method. The orientation was first determined from a Laüe X-ray back reflection photograph, and the foil was then deformed in stages until slip bands were visible on the surface under the optical microscope. This corresponded to a strain of about 4%. In the thin films prepared from this foil the major obstacles to dislocation movement seemed to be low-angle sub-boundaries, and dislocations could be observed in the process of interacting with and breaking through these boundaries (Fig.8). Well developed planar networks were present and it was possible to determine their orientations relative to those of the slip bands on the surface (Figs.9 and 10). The evidence for non-crystallographic slip in this foil is derived from the following effects:

TABLE I Dislocation density and sub-grain misorientation

Grain size	Elongation, %	Flow stress, tons/in²	Average misorientation across sub-grain boundaries	Dislocation density, cm⁻²
Annealed	...	...	...	10^6–10^7
Quenched from 850°C	...	...	...	5×10^8
50 μ	10	10	1–2°	10^{10}
80 μ	3	9	0·25°	4×10^9
80 μ	7	11	½–1°	3×10^{10}
160 μ	7·5	8	½°	3×10^9
160 μ	9	6·5	...	2×10^9
160 μ	12·5	7	½°	4×10^9
1·87 mm	4	3·5	0·3°	2×10^8
1·87 mm	12	6	...	5×10^8
Decarburized 90 μ	0·5	2·7	...	10^7–10^8
Decarburized 90 μ	1·2	4·24	...	1×10^8
Single crystal foil	3·7	5·24	0·25°	3×10^8



9 *Same specimen as Fig.8, dislocation networks showing threefold and fourfold nodes bearing no apparent relation to direction of slip traces observed on surface. Note small dislocation loops at A. The dislocations at B show abnormal dotted contrast* ×40000



10 *As Fig.9. Dislocation loops are visible at A* ×40000

Laüe asterisms

The asterism of the Laüe spots after deformation does not indicate that the misorientation in the lattice is occurring about any one specific direction (Fig.11). This may be compared with results obtained by Maddin and Chen[12] on molybdenum and niobium in which the axes of misorientation are such as to suggest strongly that slip is confined to the {110} planes.

Slip bands on the surface

The slip bands on the surface are of the diffuse, wavy type commonly observed in polycrystals. It has been deduced from the examination of replicas from the surface that if crystallographic slip is responsible for this type of structure then the segments of the slip line must be less than 30 Å long.

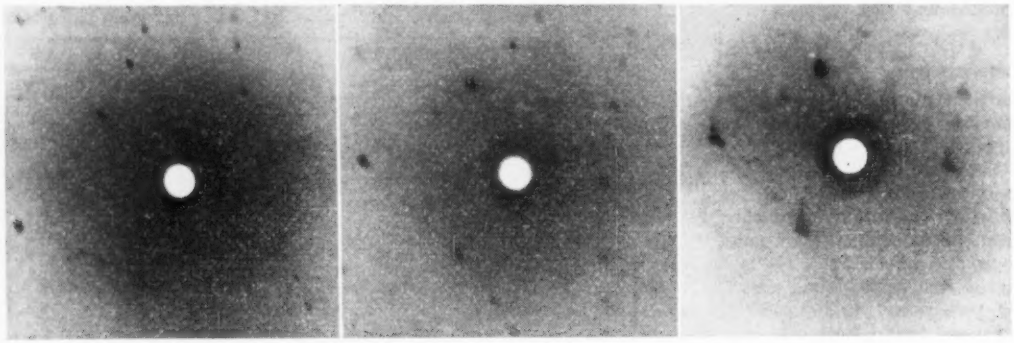
Dislocation arrangements

As in polycrystalline foils, the <100> dislocations do not lie along any specific crystallographic direction (Figs.9 and 10).

Although the techniques used for studying this single crystal foil did not enable the position of a slip band on the surface to be related to a particular position within the foil, it has been observed that in some areas large numbers of small dislocation loops are visible (*A*, Figs.9 and 10). It might be expected that since the slip bands are regions through which large numbers of dislocations have passed, they will also be regions in which large numbers of point defects will be formed during deformation. The condensation of vacancy clusters to give small dislocation loops has been observed in quenched aluminium,[13] and similar effects may account for loops forming in the wake of a slip band.

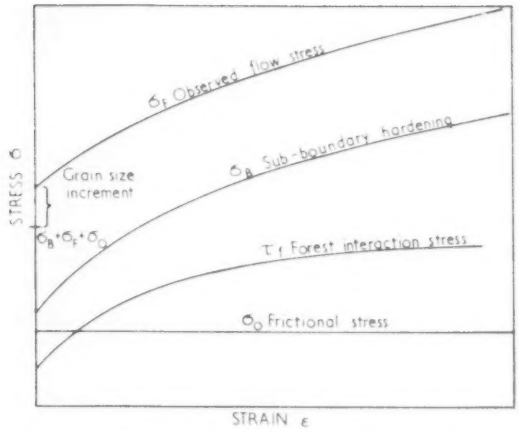
DISCUSSION

The absence of regular pile-ups of dislocations in deformed α-iron is not altogether unexpected since cross slip is known to occur readily, and even in a face-centred cubic alloy of low stacking fault energy, e.g. stainless steel, classical pile-ups are only observed at the very early stages of deformation.[3] However, the observed dependence of the lower yield stress upon grain size



a As annealed *b* After 2% strain *c* After 4% strain

11 *Same specimen as Figs.8–10 showing development of Laue asterism during deformation and absence of any definite misorientation axis*



12 *Schematic diagram illustrating factors contributing to work hardening in iron*

has been explained by Petch on the assumption that yielding is propagated by internal stresses due to classical pile-ups.[14]

Petch has found a relationship of the type

$$\sigma_y=\sigma_o+kD^{-\frac{1}{2}} \quad \ldots\ldots(1)$$

where σ_y=yield stress, σ_o=frictional stress, k=constant depending upon G (the shear modulus), b (the Burgers vector), and τ_B the pinning stress which has to be overcome to initiate yielding; D=the grain diameter.

Equation (1) can be rewritten in the form

$$\sigma_y=\sigma_o+k'(\tau_B/D)^{\frac{1}{2}} \quad \ldots\ldots(2)$$

and the term $k'D^{-\frac{1}{2}}$ may then be looked upon as a stress concentration factor, and its dependence upon grain size may be established by assuming that a dislocation source in the centre of a grain operates until the back stress at the source from a piled-up group of dislocation at the boundary is equal to $\sigma_y-\sigma_o$.

The absence of classical pile-ups in α-iron, as observed in the form of thin foil in the electron microscope, has to be reconciled with the excellent agreement between the experimental results and theoretical analysis of Petch and Heslop.[15] It may be that classical pile-ups are instantaneously formed but that they subsequently relax by a variety of possible mechanisms. Alternatively it is possible that the irregular pile-ups actually observed produce stress concentration effects which are similar to those to be expected from a classical pile-up. The absence of sharp slip lines in iron[11] strongly suggests that linear pile-ups are never formed, therefore the second explanation seems the more likely.

If an irregular pile-up is formed, dislocation sources will still operate until the back stress on the source is equal to the stress required to unpin sources in neighbouring grains, but the number of dislocations emitted by the source will be greater than if a classical pile-up were formed, i.e. the yield point extension will be greater.

Ball[5] has adapted the Petch equation to account for work hardening by assuming that the barriers to dislocation movement are the sub-grain boundaries formed during deformation. Ball now finds a relationship of the type

$$\sigma=\sigma_o+k(\tau_B/t)^{\frac{1}{2}} \quad \ldots\ldots(3)$$

where σ=flow stress, σ_o=frictional stress, k=constant, τ_B=barrier strength, t=sub-grain diameter.

Petch's calculated value for τ_B is G/15 which is close to the value of G/13 given by Cottrell[16] as the stress necessary to unpin a dislocation from a carbon atmosphere in the absence of thermal activation. The barrier stress obtained by Ball from equation (3) was G/75. This reduced value of τ_B is easy to understand since the dislocation sources operating after the yield point extension has been exceeded do not have to break free from impurity atmospheres. It is not easy to understand why τ_B is constant during the progress of deformation, neither is it to be expected that σ_o, the frictional stress, should remain constant. Cracknell and Petch[17] have shown that σ_o depends upon: (*a*) the Peierls stress, (*b*) the presence of point defects in the lattice, (*c*) the presence of segregates or precipitates, (*d*) the interaction of the dislocations with other dislocations on different slip systems.

Although the Peierls stress will not change during plastic deformation, point defects are introduced, segregates are dispersed, and the dislocation density increases: therefore it is not to be expected that equation (3) can adequately represent work-hardening behaviour.

It is probable that the greatest change in σ_o will occur as a result of dislocation interactions which are equivalent to forest hardening. Friedel[18] has shown that forest hardening should obey an equation of the type

$$\tau_f=\chi Gb/l \quad \ldots\ldots(4)$$

where τ_f=shear stress required to force a dislocation through the forest, χ=constant having a value ~0·2, l=density of forest dislocations.

Hirsch[19] has pointed out that dislocation interactions during forest hardening will obey a range of force laws depending upon the relative orientation of the two dislocation lines and their Burgers vectors, thus Friedel's equation (4) only represents an average value of the hardening stress.

Using the observed values of the dislocation density within the sub-grains, it can be shown that at the early stages of plastic deformation of decarburized foils, forest hardening accounts for a maximum of one-half of the observed flow stress and with larger deformations this contribution falls to a maximum of one-fifth.

With coarse-grained decarburized iron it now seems reasonable to describe work hardening in the following terms. Initially the dislocations move freely to the grain boundaries and form irregular pile-ups. With increasing deformation, dislocations on other slip systems become active and this results in dislocation interaction giving forest hardening, but at the same time the dislocations are reacting to form sub-boundary networks of slightly differing orientation. These networks form stronger barriers to dislocation movement than those produced during forest interaction, consequently the formation of the networks will bring about a further increase in the flow stress. As the dislocation density within the sub-grains does not vary appreciably with deformation, once the sub-boundaries have formed, the contribution of forest interaction

to flow stress will remain constant. The observed flow stress will be the sum of these various contributing factors as shown schematically in Fig.12.

In apparent contradiction to this mechanism of work hardening, Hyam and Nutting[6] and Irvine and Pickering[7] have shown that the hardening of normalized and quenched and tempered steel is inversely proportional to the square root of the ferrite grain diameters. This can only be explained if the contribution of sub-boundary hardening to the flow stress is relatively small. But with the tempered steel the grain sizes (1–10 μ) were smaller than those used in the present investigation (50 μ–2 mm). As sub-grain sizes smaller than 1–2 μ have not been observed in these experiments, even at large strains, it might be expected that as the grain size approached the sub-grain size, the contribution of sub-boundary hardening to the flow stress would become less, and in fine-grained ferrite work hardening will arise chiefly from forest hardening.

ACKNOWLEDGMENTS

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The effect of heat-treatment and microstructure on the high-temperature ductility of 18%Cr–12%Ni–1%Nb steels

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INTRODUCTION

AUSTENITIC STAINLESS STEELS containing niobium have been extensively used in the UK and the USA[1] for high-temperature service in steam power plant. In both countries difficulties have arisen during service. These difficulties have been attributed to the fact that.

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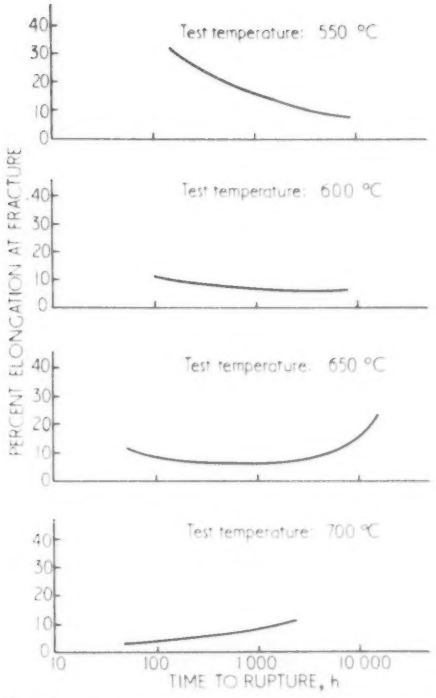
SYNOPSIS

An investigation has been made of the effect of heat-treatment and strain on carbide precipitation processes so as to correlate the resulting microstructures with the properties of an austenitic stainless steel containing niobium. It was found that NbC *can precipitate rapidly from the matrix under conditions of plastic strain. Loss of ductility in rupture and bend test specimens was always accompanied by a fine precipitate of* NbC *which hardened the matrix, causing the grains to resist creep deformation better, and throwing plastic deformation on to the grain boundaries. Grain boundary sliding and intergranular fracture resulted, and ductility declined sharply. Ductility can be improved by ageing at 900°C after solution treatment.* 1850



1 *Effect of testing temperature on relationship between ductility and rupture life. Solution treatment temperature 1050 C*

while the steels have high rupture strength, they tend to have low rupture ductilities after long testing periods, and this is invariably associated with an intergranular type of failure. On the other hand, many actual failures have occurred in the heat-affected zones of highly restrained welds, and an explanation which seems to be favoured in American work[2] is that grain-boundary liquation is responsible for the low ductility of this region.

When 18–12–1 Nb steel was selected for power plant service in this country, the high-temperature data related only to the creep strength which was very good. At about that time, however, the importance of creep ductility was becoming more widely recognized, and rupture test programmes were initiated. Some of the British data was published[1] in 1957, when it became apparent that the steel could exhibit a low level of creep ductility.

It was also found that low levels of ductility were exhibited during hot bending this type of steel at temperatures which were dependent upon the solution treatment conditions. A consideration of both this behaviour and the weld cracking in the heat-affected zones, led to the suggestion that the cause of the trouble might be the heating conditions in the heat-affected zone which produced carbide precipitation and a reduced ductility.

This aspect of the problem, namely carbide precipitation and reduced ductility, appeared to be a common feature to the three examples of low ductility described, especially as much precipitation can be accentuated by straining. It was decided therefore to investigate the mechanism of low ductility more closely. The main object of this investigation was to study the effect of heat-treatment and strain on the carbide precipitation processes and to correlate the resulting microstructures with the properties of an austenitic stainless steel containing niobium.

TABLE I Analyses of steels

Steel no.	1	2	3	4	5	6
Origin of cast	Brown Firth	United Steels	United Steels	United Steels	United Steels	United Steels
Tests carried out	Rupture	Rupture	Rupture	Bend	Tensile	Solubility of NbC
Analysis, %						
C	0·12	0·07	0·06	0·08	0·08	0·07
Mn	1·41	1·31	1·21	1·53	1·20	1·34
Si	0·78	0·37	0·31	0·53	0·41	0·42
Cr	18·46	18·22	12·98	18·14	18·80	16·80
Ni	12·13	12·55	13·09	12·64	11·80	13·25
Nb	1·33	0·93	0·89	0·95	0·80	0·94

EXPERIMENTAL

Three methods were used to investigate the ductility phenomena:

(i) an examination was carried out on a series of specimens which had been solution treated at 1050 C, and rupture tested at temperatures between 550° and 700 C for times up to 30000 h

(ii) an examination was carried out on specimens which were solution treated at various temperatures and then bent at different temperatures in the range 560–950 C

(iii) elevated temperature tensile tests were also carried out at various temperatures after different solution treatment conditions to establish the effect of both solution treatment temperature and testing temperature on the tensile ductility.

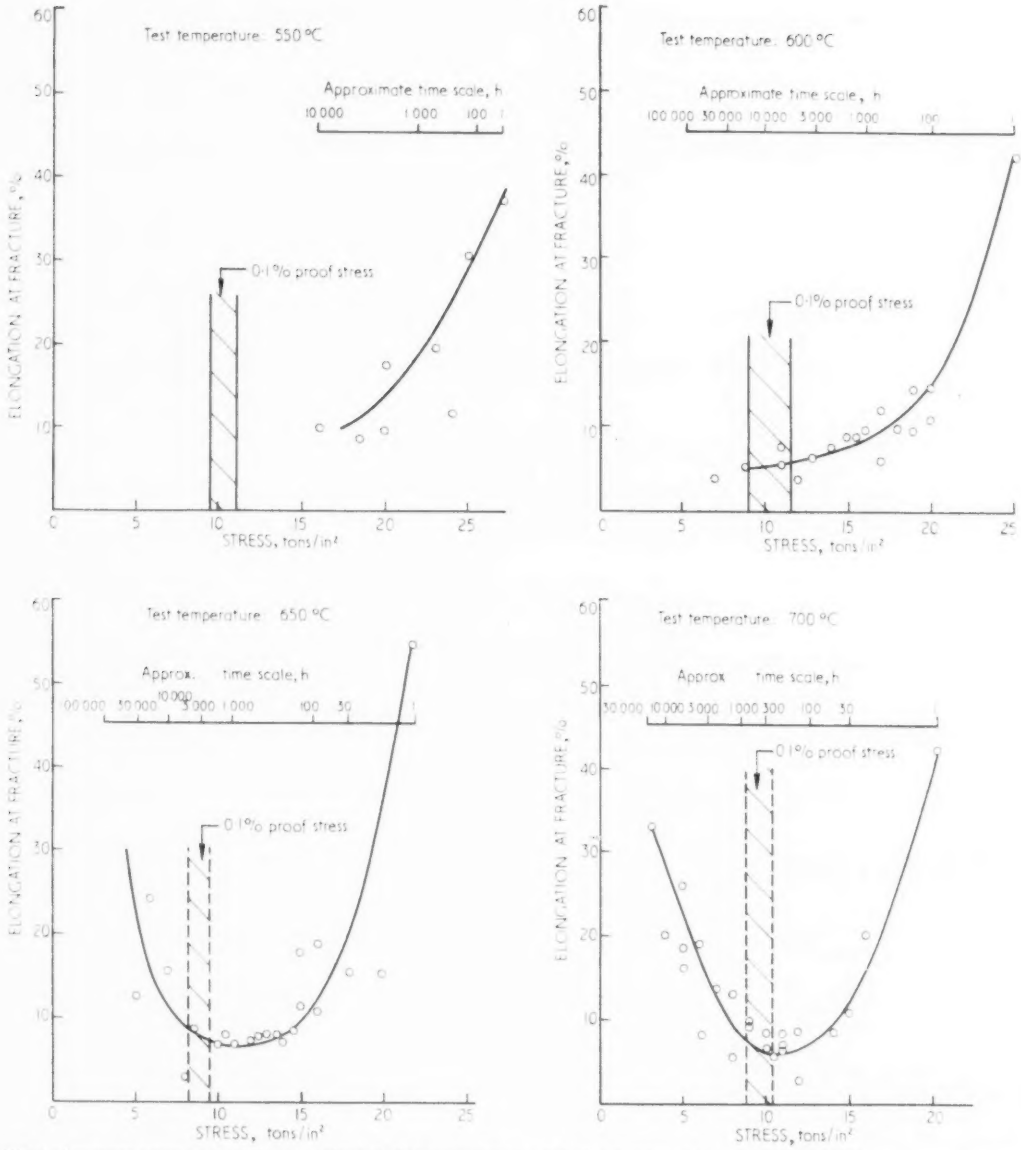
A selection of specimens from both the rupture tests and the bend tests was then examined metallographically using both the optical and the electron microscope, so that the structural changes in the specimens could be correlated with the ductility values obtained. Carbon extraction replicas were used for the electron microscopical work, and the identification of the precipitating phases was carried out using both electron diffraction and X-ray diffraction techniques.

RESULTS

Rupture tests

Ductility measurements

Tensile and rupture tests were carried out on three casts of material, the compositions of which are given in Table I. Specimens were solution treated at 1050 C and rupture tested at temperatures between 550° and 700 C using stress values between 3 and 27 tons/in². A tensile test was also carried out at each temperature. The elongation at fracture and the rupture life are given in Table II. The conventional method of presenting creep ductility data is to show the elongation at fracture as a function of the rupture life. Such a method applied to 18–12–1 Nb steels shows that with increasing rupture time the creep ductility falls to a minimum and then recovers. The time to reach the minimum ductility decreases as the test temperature increases. Recently, however, Glen[3–5] has published a



2 *Effect of rupture stress on ductility at various testing temperatures. Solution treatment temperature 1050°C*

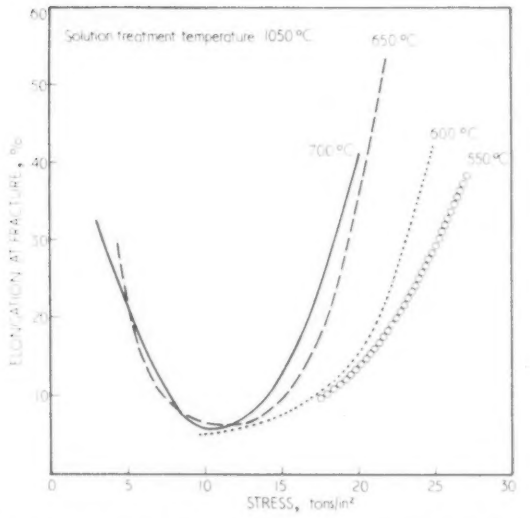
series of papers on a new approach to creep problems. One of these papers[5] considered creep ductility, and showed that the elongation at fracture could be expressed as a function of stress rather than of rupture time. As will be shown later, this method has given an interesting clue to the mechanism of the low creep ductility.

The ductility values are plotted as a function of the rupture time in the conventional manner in Fig.1 from which it can be seen that the ductility minima occur at shorter times as the testing temperature increases. The graphs in which the data are plotted, in the manner shown by Glen, are presented in Fig.2, and it can be seen that the stress which produces the minimum ductility approximates very closely to the 0·1% proof stress in a tensile test carried out at the same temperature. The mean curves for all the temperatures between 550° and 700°C are superimposed upon one another in Fig.3, which shows that the minimum elongation is about the same at each test temperature.

Microstructures

Specimens were examined from various positions on the curves shown in Figs.1 and 2. The work was concentrated mainly on specimens tested at 650°C, but sufficient specimens were examined at other test temperatures to confirm the general effects which were observed. An examination of the optical microstructures showed no effects which could be related to the ductility results, and attention was concentrated on the structures revealed by electron microscopy.

Tests at 550°C The electron microstructures are



3 *Effect of testing temperature on relationship between ductility and rupture stress*

shown in Fig.4. The tensile specimen showed some undissolved NbC but little or no precipitation (Fig.4*a*) because of the short time at temperature. In the rupture test specimens, a stress of 25 tons/in² produced failure in 206 h with an elongation of 30·5%, but in this specimen there was no precipitation, and the structure was similar to that of the tensile test specimen. Decreasing the stress to 24 tons/in² caused rupture in 429 h with a decrease in the elongation to 19·5% and the structure now showed a fine precipitate of NbC in the austenite (Fig.4*b*). Still further decreasing the rupture stress to 18 tons/in² resulted in a life of 3176 h, and the ductility had decreased to 8·5%. There were now very heavy precipitates of NbC in the austenite (Fig.4*c*), and there was also considerable evidence of intergranular cracking as shown in Fig.4*d*.

Tests at 600°C A similar series of structures was observed with decreasing stress or increasing rupture life, as shown in Fig.5. It can be seen that as the rupture life increased the ductility markedly decreased. Associated with this decrease in ductility, a fine precipitate of NbC was observed in the matrix. Some $M_{23}C_6$ carbide was also identified on the grain boundaries, and in some places, especially around undissolved NbC, there was preferential precipitation of NbC on dislocation lines.

Tests at 650°C The structures which were observed at this test temperature are shown in Fig.6. As the rupture stress decreased and the time to rupture increased, the ductility approached a minimum, and this reduction in ductility was accompanied by the formation of small NbC precipitates within the matrix austenite.

The tensile specimen showed no precipitation of small NbC particles (Fig.6*a*) and the structure showed undissolved NbC particles in the matrix. Decreasing the rupture stress to 16 tons/in² reduced the ductility to 10·5%, and fracture occurred in 90 h. The structure showed very many extremely fine NbC precipitates throughout the grains (Fig.6*b*). With a rupture stress of 14·5 tons/in² the ductility had further decreased to

TABLE II Details of rupture test specimens used for microscopical examination

Type of test*	Steel no.	Temp., °C	Stress, tons/in²	Life, h	Elong. at fracture, %
T	2	550	27·1	≏0·5	37·0
R	2	550	25	206	30·5
R	2	550	24	429	19·5
R	2	550	18	7176	8·5
T	2	600	25·2	≏0·5	42·0
R	2	600	19	129	9·5
R	2	600	15·5	663	8·6
R	2	600	14	2430	7·5
R	3	600	12	5318	3·5
R	3	600	11	6661	5·2
T	2	650	22·5	≏0·5	54·0
R	2	650	16	58	10·5
R	2	650	14·5	154	8·2
R	2	650	13·5	192	7·7
R	2	650	12	487	7·2
R	2	650	11	964	6·5
R	1	650	8·5	3900	8·5
R	1	650	7	10970	15·4
R	1	650	6	14923	22·9
R	3	650	5	34704	12·5
T	3	700	20	≏0·5	42·0
R	3	700	12	48	2·8
R	3	700	10·5	119	5·8
R	3	700	10	165	6·5
R	3	700	6	1834	8·0

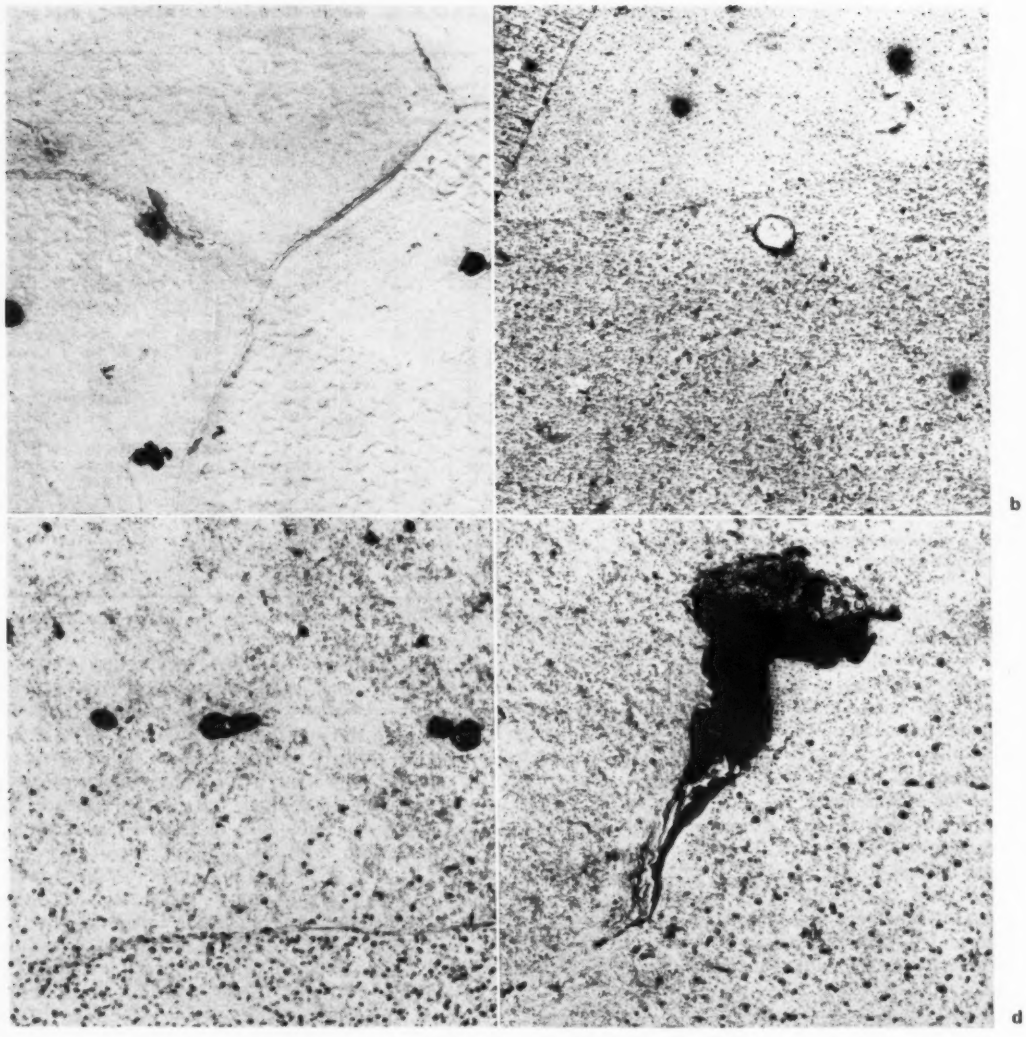
*T = tensile R = rupture

8·2% and there were now very many but slightly larger NbC precipitates in the structure (Fig.6*c*). As the rupture stress was decreased further to 13·5 and 12 tons/in², the ductility decreased to 7·7% and 7·2% while the rupture life increased to 487 h at a stress of 12 tons/in². The structures all showed fine precipitates of NbC (Figs 6*d* and *e*), but as the rupture life increased, the particle size of the NbC also increased. The minimum ductility of 6·5% was observed at a rupture stress of 11 tons/in² which produced failure in 964 h, and the structure was similar to that shown in Fig.6*e*. With a further decrease in the rupture stress, the ductility began to improve and the structures in Figs.6*f*, *g*, and *h* show that the precipitated NbC particles were very much larger, and fewer in number. At a stress of 6 tons/in² the ductility had recovered to a value of 22·9%.

As can be seen, at this temperature there was considerably more NbC precipitated, and $M_{23}C_6$ carbides were also present at the austenite grain boundaries and also along twin boundaries. Intergranular fracture was also observed, together with some denudation at the grain boundaries, especially in specimens showing very reduced ductility.

Tests at 700°C At this temperature only two specimens were examined, at stresses of 12 and 6 tons/in². The higher stress of 12 tons/in² produced a very low ductility of 2·8%, and the microstructure showed a very fine precipitate of NbC, together with preferential precipitation at sub-boundaries, and denudation effects around the austenite grain boundaries (Fig.7*a*). Decreasing the stress to 6 tons/in² resulted in an increase in ductility to 8·0%, and the NbC precipitates were very much coarser (Fig.7*b*).

It can be seen from the graphs of ductility against rupture stress in Fig.2 that the minimum ductilities were associated with stresses around the 0·1% proof stress values at the temperature of testing. It is suggested, therefore, that at stresses above the 0·1%



a Tensile specimen 37·0% elongation *b* 24 tons/in² 19·5% elongation *c* 18 tons/in² 8·5% elongation *d* 18 tons/in² 8·5% elongation, showing intergranular fissure

4 *Microstructures of specimens, rupture tested at 550°C* ×15000

proof stress, some deformation occurs during loading, and leads to strain-induced precipitation of NbC, which is responsible for the reduced ductility.

Hardness values

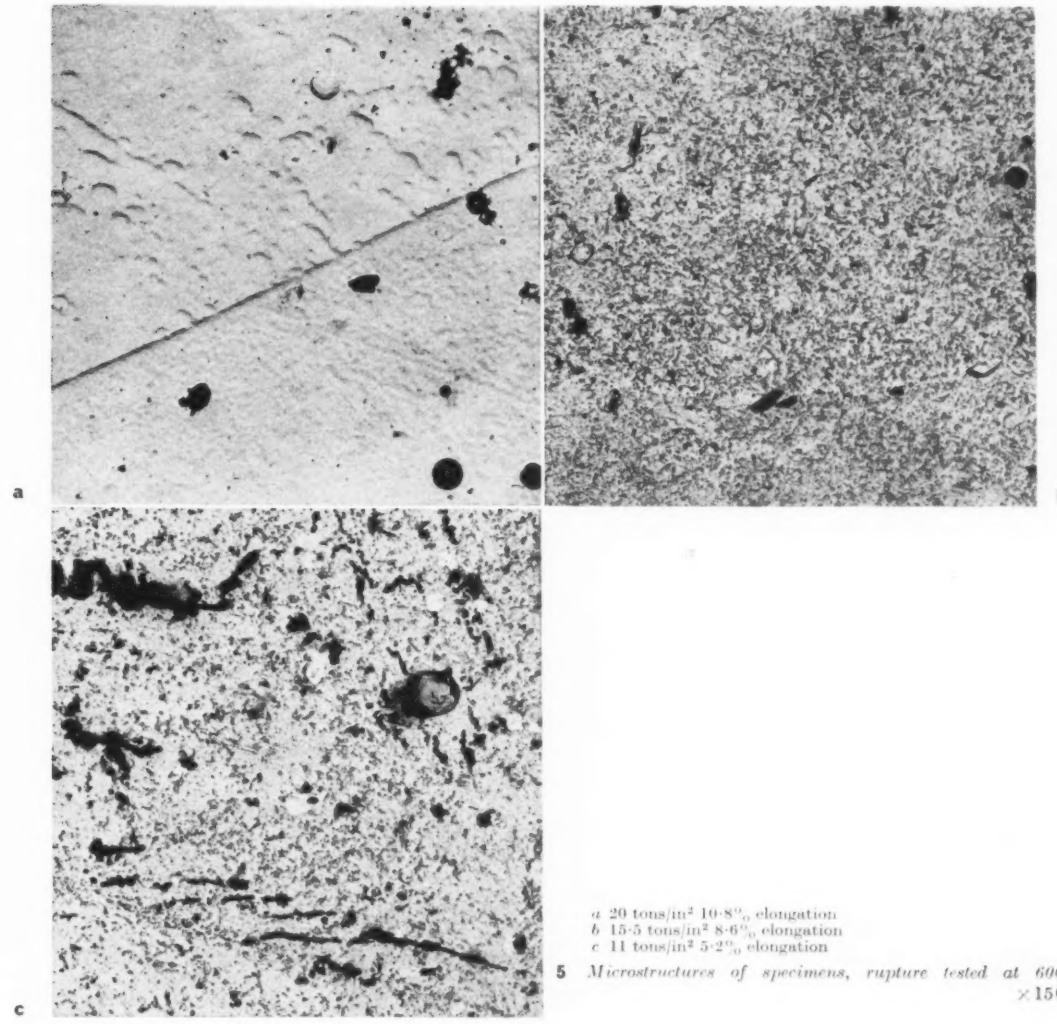
Hardness values were determined at positions near to, and remote from, the fracture of the rupture specimens. While the results were not very conclusive, they did indicate that when there was a decreased ductility, the regions near to the fracture tended to be slightly harder. This may have been due to strain-induced precipitation. When the ductility recovered at stresses below that of the 0·1% proof stress the regions near to the fracture tended to be slightly softer.

Bend tests

Ductility measurements

Specimens from steel no.4 were solution treated at varying temperatures between 1050° and 1300°C and bent at temperatures between 800° and 950°C. The bend test was completed in under 5 min so that the length of time at temperature was short. The ductility of these specimens was measured by the bend angle when cracking began, and the results are shown as a function of both testing temperature and solution treatment temperature in Fig.8. It can be seen that a loss of ductility occurred with solution treatments above 1100°C during bending at 800° and 825°C, but rather higher solution treatment temperatures, of the order of 1150°C were required for the ductility to be decreased at a bend temperature of 875°C. Using bend temperatures of 950°C a very high solution treatment temperature of the order of 1300°C was required to cause the ductility to decrease.

The effect of the bend temperature on the ductility when using a solution treatment temperature of 1150°C is shown in Fig.9. It can be seen that as the bend temperature increased, the ductility decreased,



a 20 tons/in² 10·8% elongation
b 15·5 tons/in² 8·6% elongation
c 11 tons/in² 5·2% elongation

5 *Microstructures of specimens, rupture tested at 600°C* ×15000

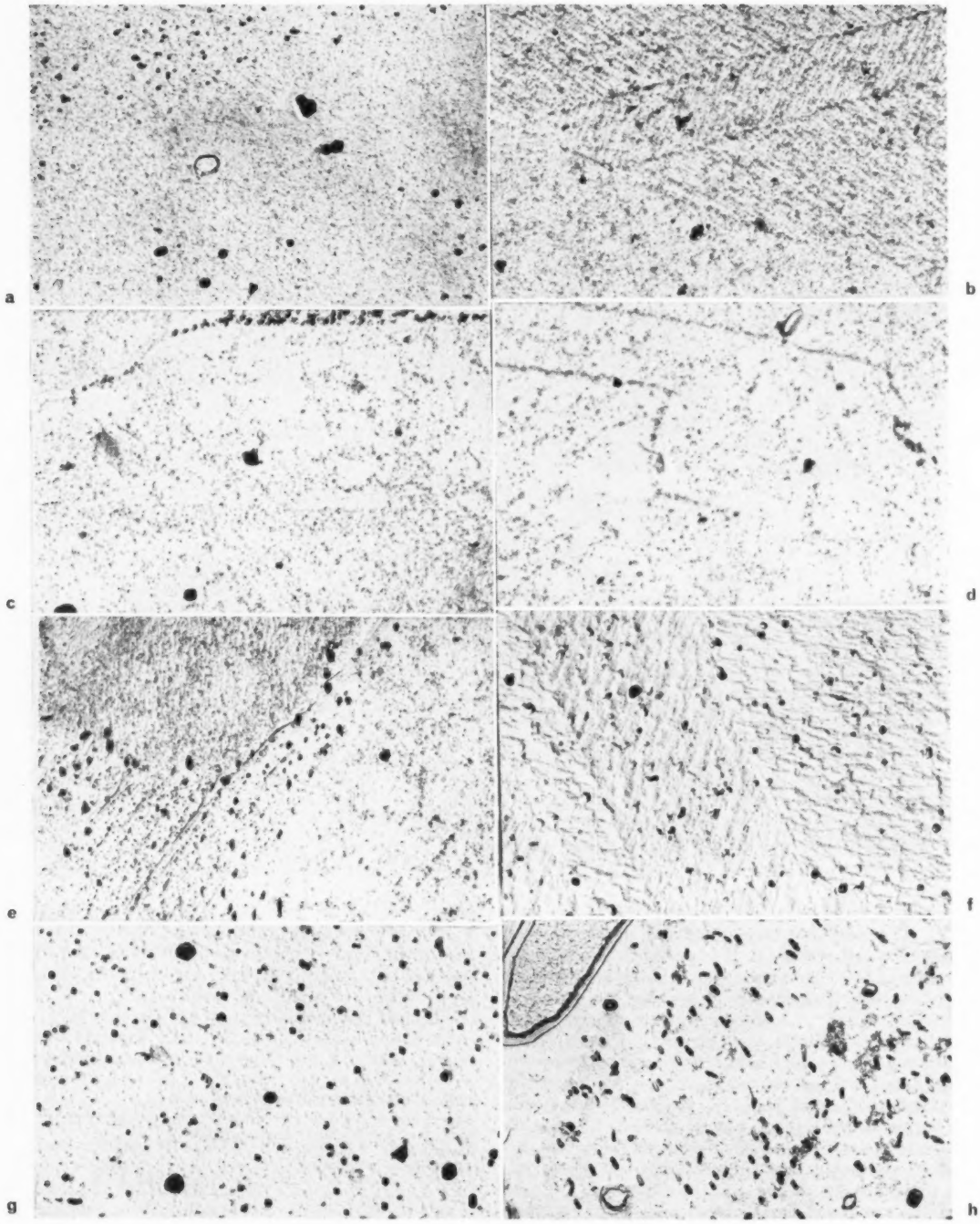
and reached a minimum in the temperature range 800–825°C. The ductility then increased again at bend temperatures above 850°C. If a specimen was solution-treated at 1150°C and then aged for 1 h at 900°C, no loss of ductility was observed even at a test temperature of 800°C at which the minimum ductility was observed in solution treated material. Details of the testing conditions, the bend angle, and the hardness in the strained and unstrained regions of the bend test specimen are given in Table III. It can be seen that in this series of tests there was a marked relationship between the difference in hardness between the strained and unstrained regions of the specimen (ΔH) and the ductility. When the ductility was low, the specimen was much harder near to the fracture, than remote from the fracture, but when the ductility was high, there was much less difference between the hardness of the strained and unstrained material. This relationship is clearly shown in Fig.9.

Microstructural changes

Electron micrographs from the strained and unstrained regions of the bend test specimens are shown in Fig.10. After bending at 600°C and 650°C, reasonable ductilities were obtained, and the structures of the strained and unstrained regions were identical and no precipitation of NbC had occurred (Fig.10*a*). With bending at 700°C, the ductility had been considerably reduced and it can be seen (Fig.10*b*), that the strained region showed a very fine precipitation of NbC, but

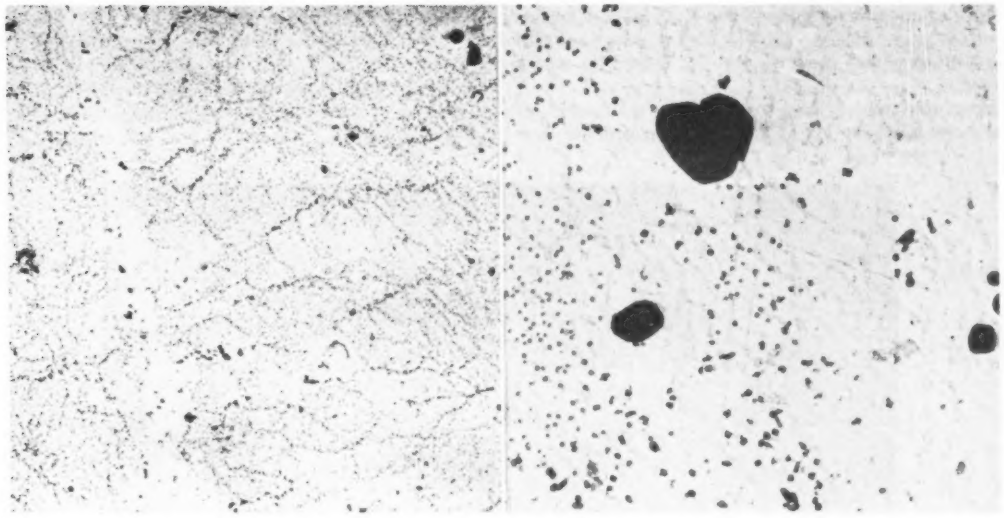
TABLE III Details of bend tests

Solution treatment temp., °C	Bend test temp., °C	Hardness DPN		ΔH, DPN	Bend angle, degrees
		Strained	Unstrained		
1150	560	...	...	...	135
1150	650	199	164	35	118
1150	700	243	167	76	80
1150	750	236	161	75	67
1150	800	...	...	...	70
1150	825	229	146	83	60
1150	850	243	180	63	160
1150	950	184	159	25	180
1300	950	207	165	42	65
1200+1 h 900	800	194	166	28	180



a Tensile test 54% elongation *b* 16 tons/in² 10·5% elongation *c* 14·5 tons/in² 8·2% elongation *d* 13·5 tons/in² 7·7% elongation
e 12 tons/in² 7·2% elongation *f* 8·5 tons/in² 8·5% elongation *g* 7 tons/in² 15·4% elongation *h* 6 tons/in² 22·9% elongation

6 *Microstructures of specimens, rupture tested at 650 C* ×15000



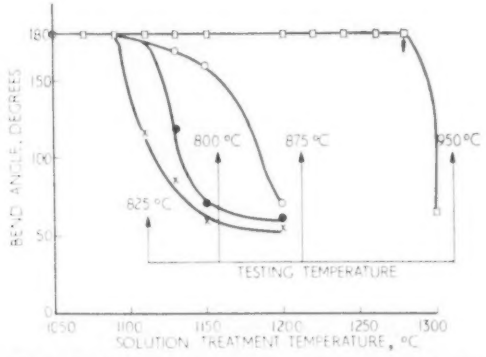
a 12 tons/in² 2·8% elongation × 3 000 *b* 6 tons/in² 8% elongation × 15 000

7 *Microstructures of specimens, rupture tested at 700°C*

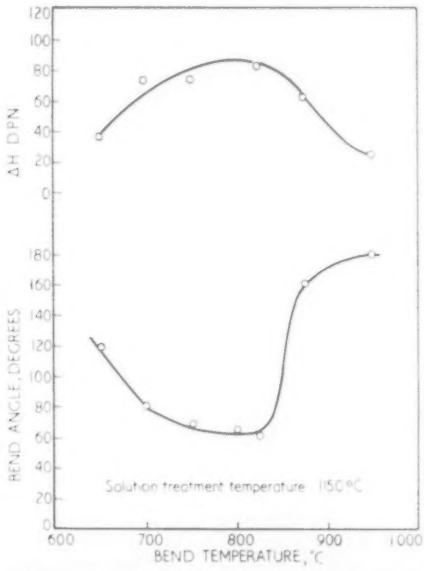
there was no precipitate in the unstrained region (Fig.10*c*). A similar effect of precipitation of NbC in the strained region, but not in the unstrained region (Figs.10*d* and *e*), was observed after bending at 750°C, at which temperature the bend angle had decreased to 67°. The lowest ductility was obtained by bending at 825°C, and while there was a little precipitation of NbC in the matrix of the unstrained region (Fig.10*f*) the strained region showed an extremely heavy precipitate of NbC, together with marked denudation effects at the grain boundaries (Fig.10*g*). Higher ductility was obtained during bending at 875°C, and while the unstrained structure showed some precipitation of NbC, a much heavier precipitation was observed in the strained regions (Fig.10*h*). However, the particle size of the precipitates was greater in the specimen bent at 875°C compared with the specimens bent at lower temperatures. Ductile bends were obtained at 950°C, and the structures of both the unstrained and strained regions were identical and showed fairly large precipitates of NbC (Figs.10*i* and *j*). These carbides were very much coarser than those produced at lower bending temperatures. Preferential precipitation was often observed on dislocations, especially around the undissolved NbC.

From these results it can be seen that straining induced a heavy precipitation of NbC which was associated with a decreased ductility. However, after bending at 875°C, although there was a heavy precipitate, the loss of ductility was small, due, no doubt, to the fact that the particles were much coarser. In fact, as shown in Table IV, it seems that if the particles were less than 200 Å in dia. the ductility was greatly reduced, but if the particles were larger than 200 Å the ductility was not markedly impaired.

As shown in Fig.8, specimens solution treated at 1300°C and bent at 950°C had low ductility, and the



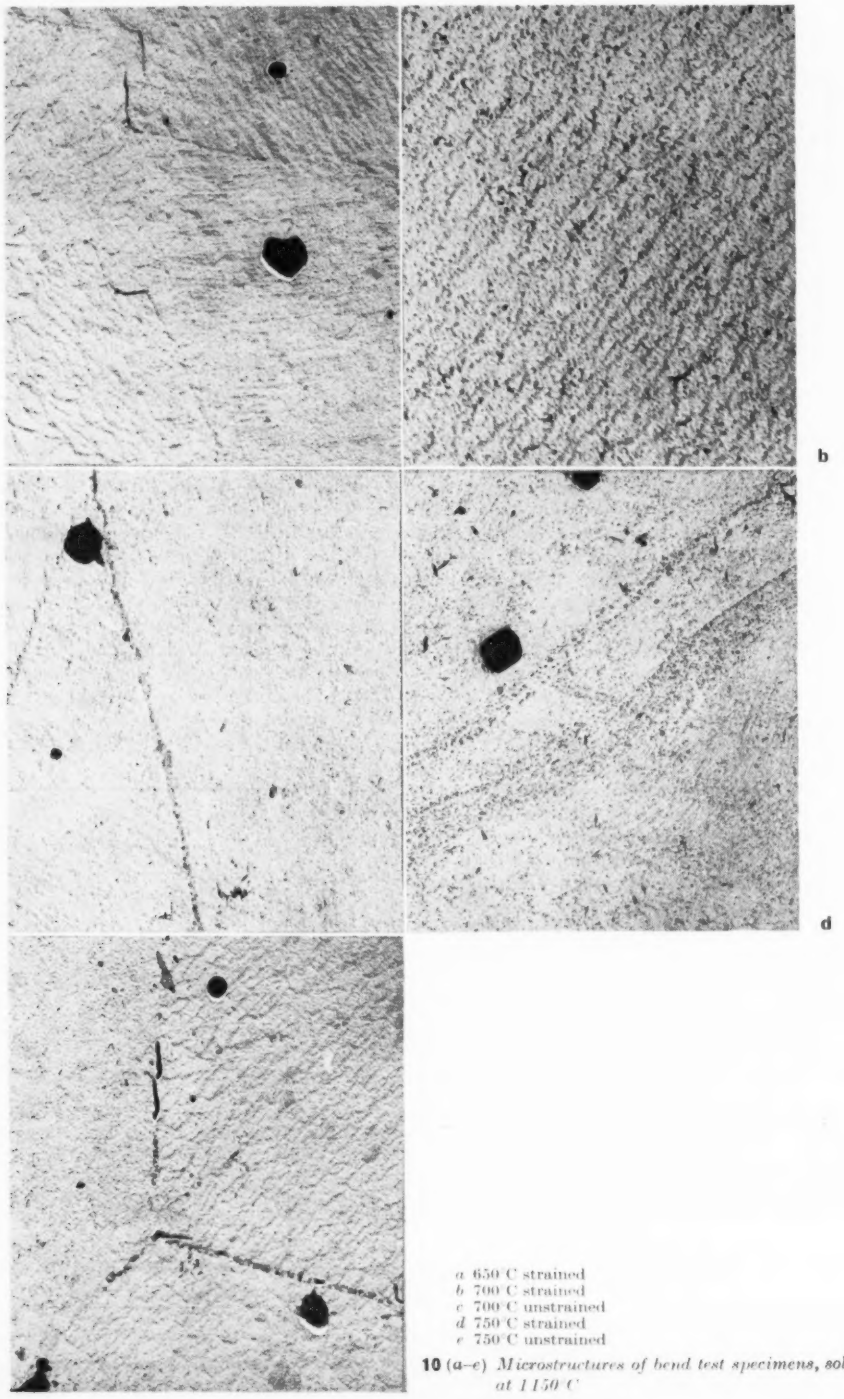
8 *Effect of solution treatment temperature on ductility during bending at various temperatures*



9 *Effect of bending temperature on ductility and hardening produced by strain-induced precipitation*

structure of specimens taken from the strained region showed fine precipitates of NbC. These were no doubt responsible for the marked loss of ductility which occurred.

A specimen which had been solution treated at 1150°C and aged for 1 h at 900°C was examined after bending at 800°C. The ductility had been completely restored. The ageing treatment had precipitated large particles of NbC, and the structures in both the strained and unstrained regions were virtually identical, consisting of large precipitates of NbC (Figs.11*a* and *b*).

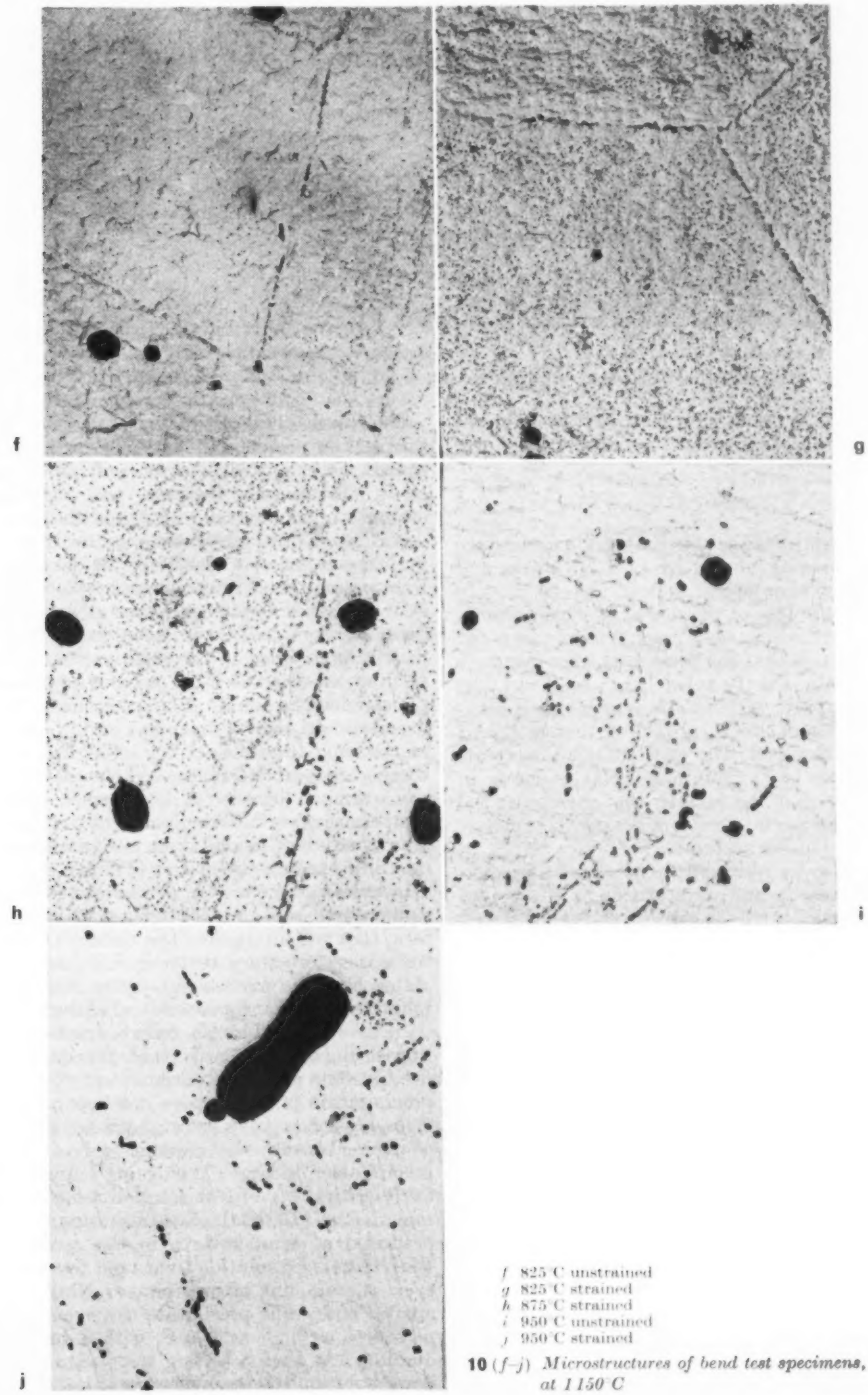


a 650°C strained
b 700°C strained
c 700°C unstrained
d 750°C strained
e 750°C unstrained

10 (*a–e*) *Microstructures of bend test specimens, solution treated at 1150°C* ×15000

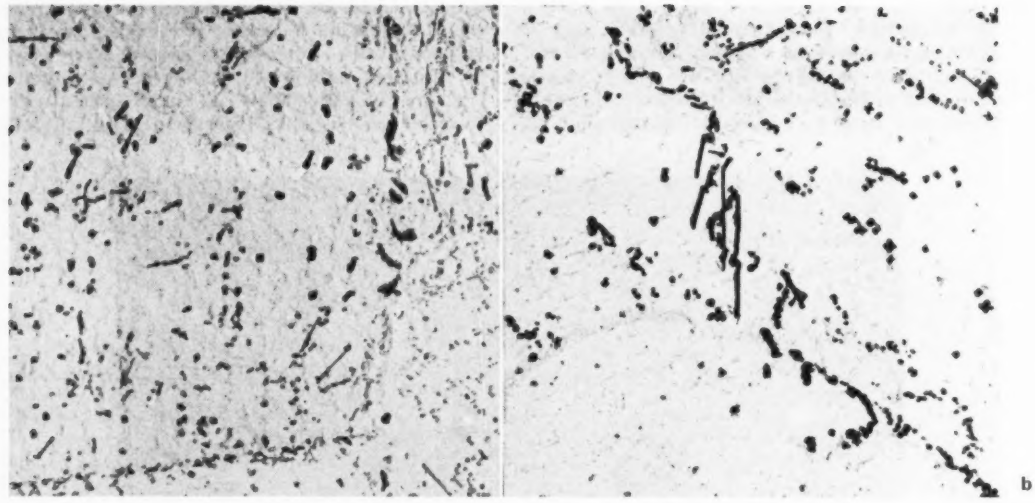
Tensile tests

Tensile tests were carried out on steel no.5 at 700–900°C, after solution treatment at 1050–1300°C. The tensile data are given in Table V. It can be seen that with increasing solution treatment temperature the tensile strength increased, and the ductility decreased, at any given testing temperature. With constant solution treatment temperature the strength decreased with increasing testing temperature, and the ductility exhibited a minimum at testing temperatures of 750–800°C. As the solution treatment temperature increased, however, the overall ductility



f 825°C unstrained
g 825°C strained
h 875°C strained
i 950°C unstrained
j 950°C strained

10 (*f–j*) *Microstructures of bend test specimens, solution treated at 1150°C* ×15000



a Bent at 800°C unstrained *b* Bent at 800°C strained

11 *Microstructures of specimen solution treatment at 1200°C. Aged 1 h at 900°C and bent at 800°C* ×15000

was lowered and the minimum occurred over a wider range of testing temperatures. These effects are shown in Fig.12.

DISCUSSION

It is clear from the above results that strain can induce rapid precipitation of NbC in 18–12–1 Nb steels, and strain-induced precipitation appears to be largely responsible for the loss of the ductility which occurs. The rapidity with which such precipitation can occur is shown by the fact that the bend tests were complete in less than 5 min. On the other hand, NbC can precipitate fairly rapidly without strain at high temperatures of the order of 950°C, providing that sufficient NbC has been dissolved. The effect of solution treatment temperature on the solution of NbC is shown in Fig.13, from which it can be seen that increasing the solution temperature markedly increases the amount of NbC dissolved in the austenite.

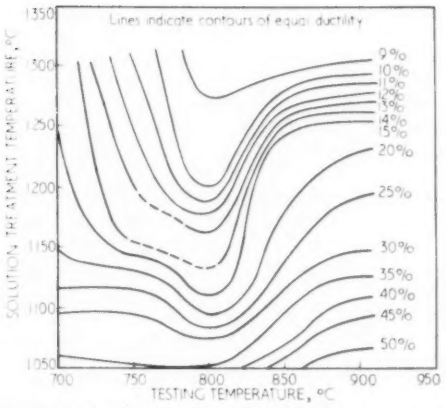
The mechanism by which strain accelerates precipitation of NbC must be one of nucleation as it increases the number rather than the size of the precipitated particles. Plastic deformation increases the dislocation density and produces a large number of possible nucleation sites. It can be clearly shown that preferred precipitation takes place on dislocations, especially when high solution temperatures have been used to produce a high degree of supersaturation (Fig.14*a*). Precipitation also occurs on sub-boundaries (Fig.14*b*), dislocation pile-ups at grain boundaries (Fig.14*c*), and dislocation arrays around undissolved NbC particles (Fig.14*d*).

TABLE IV The effect of bending temperature on the size of the precipitated NbC particles

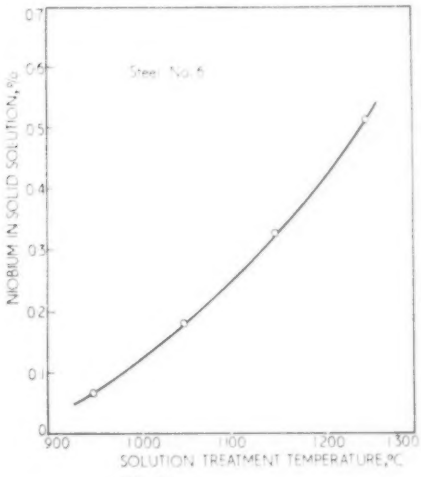
Bend test temp., °C	Approx. dia. of NbC precipitates, A°
700	60
750	60–100
825	150
875	200–300
950	400–600

The mechanism of the loss of ductility is probably due to the presence of fine NbC precipitates in the matrix, especially if they have precipitated on dislocations. These will increase the high-temperature strength of the matrix grains, and throw much of the deformation during high-temperature testing on to the grain boundaries, with the result that intergranular fracture and loss of ductility take place. This process is accentuated by the presence of a grain boundary denuded zone which is weak compared with the heavily precipitation-hardened matrix, but it must be pointed out that such grain boundary denudation is not essential for low ductilities because if the grain has been strengthened sufficiently, grain boundary deformation and loss of ductility will occur even though there is no denudation. Also, if the matrix precipitate has overaged because it has formed either at high temperatures, or for long times at low temperatures, grain boundary denudation need not result in low ductility because the matrix is now much less strong. The mere presence of a precipitate is therefore not sufficient to cause embrittlement as it must be in a form that will strengthen the matrix. Overaged structures therefore show better ductilities, and it is suggested that the particle size of the NbC must be less than 200 Å in dia. for severe loss of ductility to occur.

In the bend and tensile tests the deformation occurring during testing is responsible for the strain-induced precipitation and the associated loss of ductility. Such precipitation is much more detrimental if the specimens have first been solution treated at a high temperature, because the amount of NbC available for precipitation is large. The ageing temperature is also fairly critical because at solution temperatures in the range 1050–1150°C the fine dispersion of NbC is only produced at temperatures in the range 750–800°C. With increasing solution treatment temperature, however, the amount of precipitated NbC is very much greater and a fine precipitate can even occur at temperatures as high as 900°C, with a resulting loss of ductility. At such a testing temperature and using a lower solution treatment temperature, the precipitate



12 *Effect of solution temperature on ductility at varying tensile test temperatures*



Treatment	Nb, % Sol.	Insol.	Total
950°C, 1 h	0·065	0·88	0·945
1050°C, 1 h	0·179	0·78	0·950
1150°C, 1 h	0·325	0·62	0·945
1250°C, 1 h	0·510	0·43	0·940

13 *Effect of solution treatment temperature on solubility of niobium*

would have been both less in amount and of a larger particle size so that there would have been little or no loss of ductility. Therefore, the temperature at which there is a maximum loss of ductility will be determined by the amount of precipitate which is formed, the rate of precipitation as affected by temperature, the accelerating effect of strain, the size of the precipitated particles, and the degree to which grain boundary denudation occurs. Consequently, the temperature of minimum ductility can vary considerably as the test or service conditions are altered.

The restoration of ductility when specimens are aged at 900°C before testing at 800°C is the result of NbC precipitating as a coarse dispersion and thereby reducing the supersaturation so that little or no further precipitation can take place even under the accelerating influence of strain at lower temperatures. It is obvious that such an ageing treatment must be at a temperature at which sufficiently rapid precipitation of NbC can occur without the influence of strain, in order to reduce the supersaturation to as low a degree as possible.

In the rupture tests, the significant point is that the minimum ductility occurs at all temperatures at a stress which is close to the 0·1% proof stress. It is suggested that dislocations are generated by the plastic strain which occurs during loading, and that subsequently, NbC precipitates nucleate upon many of these dislocations and grow with increasing time of testing. It is these precipitates which are responsible for the loss of ductility. When very large amounts of plastic strain occur during testing, as in the high strain rate tests such as a tensile test, while strain-induced precipitation can take place, the time of testing is so short that at the temperatures which have been examined, i.e. 500–700°C, the amount of precipitation is very small and the ductilities are therefore relatively high. As the rupture tests become more prolonged at lower stresses, the strain-induced precipitation becomes more evident and the ductility therefore decreases. When the stress used in the rupture test is below the 0·1% proof stress, there is much less strain activation because there is no plastic deformation on loading. Consequently, there is little acceleration of precipitation during even lengthy periods of testing because the only plastic strain is the relatively small creep deformation, and what precipitation does take place becomes markedly overaged and does not decrease the ductility. Hence the ductility is restored in the specimens tested at stresses below the 0·1% proof stress.

It has already been shown that the minimum creep ductility during rupture testing occurred at increasing times, the lower the testing temperature (Fig. 1). If the process controlling this minimum in ductility is one of diffusion, then the time for the minimum ductility should be related to the temperature of testing by an equation of the type:

$$t = Ae^{-Q/RT}$$

where t = time to minimum ductility, T = absolute

TABLE V The effect of solution treatment temperature and bending temperature on the tensile properties

Bending temp., °C	700°C		750°C		800°C		850°C		890°C	
Soln. treatment temp., (all air cooled)	TS, tons/in²	Elong., %	TS, tons/in²	Elong., %	TS, tons/in²	Elong., %	TS, tons/in²	Elong., %	TS, tons/in²	Elong., %
1050	19·5	34·1	13·6	35·3	12·6	35·2	9·1	47·9	7·7	55·0
1070	20·9	33·5	16·1	33·3	12·9	33·9	9·7	40·0	9·3	43·6
1090	20·2	30·3	16·5	31·3	12·5	20·9	10·8	34·7	8·6	49·3
1110	20·9	27·5	16·8	26·9	13·4	15·3	10·4	39·4	9·5	36·3
1130	21·7	18·2	17·7	20·1	13·3	18·5	12·4	26·2	10·2	33·9
1150	21·3	20·4	18·2	13·3	14·5	16·1	11·5	17·9	8·9	28·0
1200	22·9	15·3	18·4	12·5	16·2	10·3	11·1	23·0	9·0	23·6
1250	23·8	16·1	21·5	12·3	16·6	10·3	14·5	14·3	9·7	15·1
1300	19·7	15·1	17·5	11·3	17·6	7·8	13·4	8·1	10·9	8·5

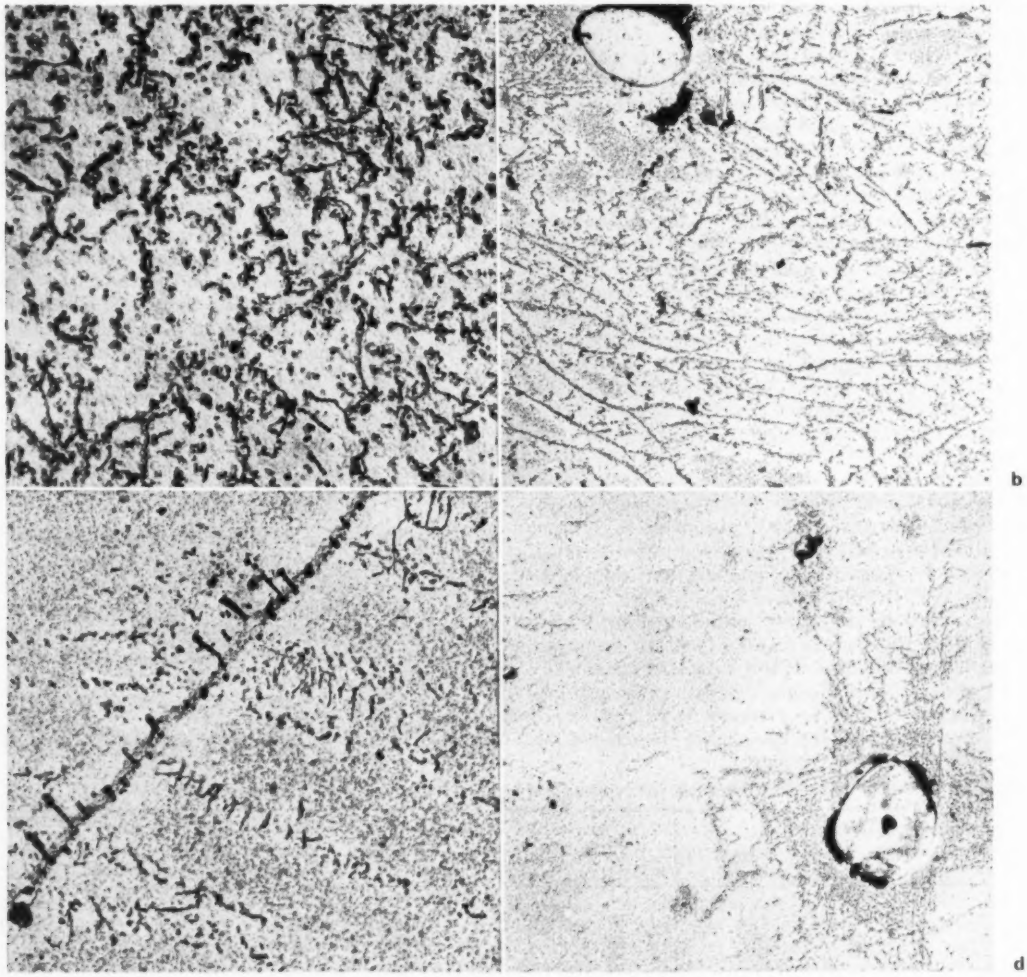
temperature, Q=activation energy for the process, A and R are constants. A graph of log t against $1/T$ should give a straight line, the slope of which will be the activation energy Q.

Sufficient data was not available to obtain the time for the minimum ductility at all the temperatures examined, but if an arbitrary low ductility value was selected, the times to reach this ductility at temperatures of 550–650°C gave a straight line graph for log t plotted against $1/T$, and the value of Q was about 18000 cal/g-mol. This was much less than the value for the diffusion of niobium in austenite, but Cottrell[6] has suggested how the measured value can be corrected for the accelerating effect of strain on the diffusion process. Using this correction, a value was obtained lying between 13300 and 18600 which is of the same order as that measured, indicating that it is the diffusion of Nb in the austenite which is the controlling mechanism for the loss of ductility.

It has already been shown that there was a marked correlation between the ductility in the bend tests and the hardness increase in the strained regions of the specimens. This hardness increase may have been partly due to some work hardening, but it is also the result of precipitation hardening. There is in fact, a marked relationship (Fig.9), between the hardening which takes place during bending and the loss of ductility, and this cannot be explained only on the basis of work hardening.

NbC precipitates often occur preferentially upon dislocations, especially after solution treatment at high temperatures. Preferential precipitation can also occur on dislocations around undissolved NbC particles (Fig.14*d*), and it has been suggested that this type of preferential precipitation results from the dislocations generated by the differences in coefficient of expansion between the NbC and the matrix.[7] Dislocations could also be formed when NbC dissolves in the austenite, and near to undissolved NbC particles the supersaturation will be high, which will contribute to forming arrays of preferential precipitates.

Preferential precipitation of NbC on dislocations would be expected to be a powerful means of strengthening the material at high temperatures when relatively



a Solution treated 1300°C. Bent at 950°C. Strained region showing precipitation of NbC on dislocations × 30000 *b* Solution treated 1300°C. Bent at 950°C. Strained region showing preferential precipitation on sub-boundaries × 3000 *c* Solution treated 1250°C. Bent at 875°C. Strained region showing precipitation of NbC on dislocation pile-ups × 15000 *d* Solution treated 1300°C. Bent at 875°C. Showing preferential precipitation on dislocations surrounding undissolved NbC particle × 3000

14 *Some examples of preferential precipitation of* NbC

few dislocations are responsible for the plastic deformation. These dislocations will be pinned by such precipitates so that the high-temperature properties will be improved. At high strain rates, however, such as in a tensile or hardness test, when very many dislocations are generated, the locking of a number of dislocations by precipitates would have less effect on the strength.

It has recently been shown that preferential precipitation can occur on dislocations during warm working,[7] and this effect is strain-activated precipitation. Such precipitation would be expected to improve the high-temperature properties and this is indeed the case, since it is well known that the creep strength is increased by warm working.

It has been clearly shown in Fig.13 that increasing the solution treatment temperature causes more niobium to dissolve in the austenite, and this results in more precipitation of NbC and a greater loss of ductility. This effect can occur in the heat-affected zone of a welded component which is under restraint, and the strain can readily activate precipitation so that a loss of ductility can result. Such may well be the explanation of cracking in highly restrained welded joints. It has also been observed that precipitation of NbC can take place during welding with 18–12–1 Nb weld metal under some conditions, and may contribute to weld metal cracking and low ductilities.

Practically, it can now be appreciated that the subsequent properties of this type of steel are very dependent on the solution treatment conditions. It is essential that there is sufficient niobium dissolved to give the requisite level of strength, but too much dissolved niobium will lead to very much poorer ductilities. The solution treatment must therefore effect a compromise between strength and ductility, but, should the tendency to lack of ductility still be too great, it can be removed by an ageing treatment at 900–950°C which will precipitate NbC in an innocuous form. The strength will then probably have to be restored by a correct solution treatment of 1050°C.

CONCLUSIONS

An examination has been made into the conditions affecting the loss of ductility in 18–12–1 Nb austenitic stainless steels, and the relationship between microstructure and the resulting properties has been established. The main conclusions which have been drawn from this work are as follows:

1. Precipitation of NbC can occur rapidly from the matrix of 18–12–1 Nb austenitic steels under the activation of plastic strain.

2. In the rupture tests, this precipitation produces a fine precipitate of NbC at temperatures of 550–700°C when the applied stress is sufficient to exceed the 0·1% proof stress, and plastic deformation occurs during loading. At stresses lower than the 0·1% proof stress, any precipitation of NbC occurs over much longer times, and is formed in a very much coarser dispersion.

3. In the bend tests, the strained regions of the specimens showed a much heavier precipitation of NbC than did the unstrained regions, and with increasing bend test temperature the size of the precipitated NbC particles became larger.

4. The loss of ductility in both rupture and bend test specimens was invariably associated with a fine general precipitate of NbC which caused hardening of the matrix so that the grains became more resistant to creep deformation and the plastic deformation was largely thrown on to the grain boundary regions. Under these conditions, grain boundary sliding and intergranular fracture resulted, with a marked loss of ductility.

5. The results of both bend and tensile tests showed that the loss of ductility was more pronounced and occurred over a wider temperature range as the solution treatment temperature increased, because of the larger amount of niobium available in solution for precipitation as NbC.

6. The conditions necessary for a decreased ductility to be observed are: that there should be fairly large amounts of NbC available for precipitation, that some plastic deformation should be applied during the precipitation process, and that the general dispersion of the precipitated NbC should cause marked strengthening of the matrix. If the NbC is overaged and does not strengthen the matrix, as occurs in very long time rupture tests, or in bend tests above 850°C, the ductility is not adversely affected.

7. While grain boundary denudation is a contributing factor to the loss of ductility, it is not an essential feature, since if the matrix precipitate is coarse and overaged, the ductility will be high despite grain boundary denudation.

8. It is possible to improve the ductility, especially in bend and tensile tests by ageing at 900°C after solution treatment, but before testing. This precipitates coarse NbC particles and lowers the amount of NbC in solution to such an extent that on subsequent testing in the normal temperature range for low ductility, there is insufficient NbC precipitated to give the effect. This prior ageing treatment, however, will lower the strength, and the material must be resolution treated before service. Such an ageing treatment, however, would enable fabrication to be carried out with less risk of cracking.

9. It has been shown that NbC can precipitate on dislocations and in that condition which is usually associated with warm working, it is believed that the optimum effect is exerted in improving the creep strength.

10. A suggestion has been made that the heat-affected zone cracking in welds of high degrees of restraint is associated with strain-induced precipitation.

ACKNOWLEDGMENTS

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Some ductility aspects of 18–12–1Nb steel

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SYNOPSIS

Weld junction zone cracking has been experienced in the UK and the USA in 18–12–1Nb steel (Type 347) weldments where the degree of mechanical restraint is high. Laboratory rupture tests across welded joints in 18–12–1Nb steel have also indicated a tendency to failure in the position where cracking in service has been encountered. In addition, cracking identical with service failures has been produced under laboratory conditions during post-welding stress-relieving treatments.

A series of short-time tensile tests at temperatures from 600–1250°C has been carried out on 18–12–1Nb steel heat-treated at high solution temperatures to simulate the thermal treatments obtained in weld heat-affected zones; two regions of low-tensile ductility, one at around 850°C and one at 1100°C, were found. The low ductility zone at 850°C is attributed to strain-induced precipitation of niobium carbide, and it is postulated that such a mechanism could explain the 'conditioning' of the weld junction in 18–12–1Nb steel, thereby rendering the steel highly susceptible to cracking. 1851

INTRODUCTION

SEVERE CRACKING TROUBLES have been experienced in steam power plants with certain 18–12–1Nb steel (Type 347) weldments which involved a high degree of mechanical restraint during the initial welding. In most cases reported,[1–4] the cracking occurred in the heat-affected zone of the parent metal and appeared to start at or near the junction between the parent metal and the weld. A typical example of the type of cracking which occurred is shown in Fig.1, which clearly indicates its intergranular nature. As far as could be judged, the cracks started from the outside surface and extended inwards as well as circumferentially (*see* Fig.2), ultimately giving rise to steam leaks.

Apparently the failures were due to creep, but some weld joints seemed to have failed in quite short times, which was not consistent with either the design stresses employed, the general conditions of service, or the known capacity of 18–12–1Nb steel to resist creep.

In reviewing the information available[5–7] on the properties of this steel, the creep and rupture properties appear reasonably reproducible, but the ductility behaviour in creep is very variable, the results showing a wide scatter. Thus, 18–12–1Nb steel can show very low creep ductility which might account for the troubles experienced in service, and when it is considered that a weld junction could behave as a notch, both factors could contribute to premature failure.

More recently, however, instances were reported[8, 9] where cracking had apparently developed during stress relieving operations after welding, and before use under service conditions.

Development of weld heat-affected zone cracking after stress relieving has also been demonstrated on experimental welds, and Fig.3 illustrates cracking very similar to that obtained on production welds, but produced in the authors' laboratory. This evidence suggested that some explanation other than poor creep ductility as ordinarily understood was responsible for failures of the type in question, and over a period of time various explanations have been presented, the current one arising principally from the work of Nippes *et al.*[10, 11] and Puzak and Rischall.[12]

These workers found that 18–12–1Nb steel exhibited a marked degree of hot shortness when tested at temperatures of 2400–2500°F (1315–1370°C) and that this low ductility persisted on cooling to 2200°F (1205°C) following initial heating to 2400–2500°F. They concluded that this property of 18–12–1Nb steel could give rise to cracking in the weld heat-affected zone in weldments involving a high degree of restraint. The low ductility region was attributed either to precipitation at grain boundaries of some phase taken into solution above 2400°F, or to the presence of actual liquid phase at the grain boundaries at these temperatures.

While the hot shortness of 18–12–1Nb steel at very high temperatures could undoubtedly contribute to the welding troubles experienced with this steel, the existence of a low ductility region between 2450°F (1340°C) and 2200°F (1205°C) does not appear to supply a complete solution to this particular problem, even though some evidence of niobium carbide fusion has been observed in the region of the weld junction by the authors. For example, the work of Nippes *et al.* does not explain why cracks occur in 18–12–1Nb weld heat-affected zones during stress relieving at 750°C. Neither does it explain why laboratory creep rupture specimens taken from 18–12–1Nb weldments, stress relieved at 750°C, fracture in the weld heat-affected zone, practically at the weld junction, when tested at 650°C, but fracture in the parent material well away

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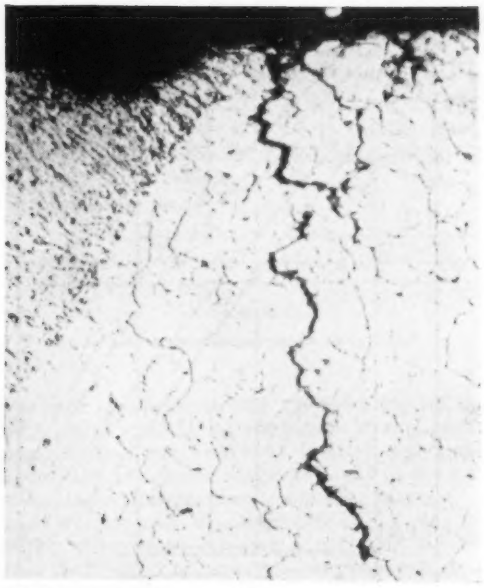


1 *Weld heat-affected zone cracking* ×100

from the weld junction if initially stress relieved at 850°C or 1050°C. In all these cases, the test pieces were examined and were free from defects before testing. The results of this latter work are given below.

Further work was instituted in an attempt to answer some of the problems outlined above.

DETAILS OF INVESTIGATION

Creep rupture tests across 18–12–1Nb welds

Before any specific investigation into weld cracking troubles with 18–12–1Nb steel, work at the Brown–Firth Research Laboratories included a programme of testing across welds in austenitic steels. Test pieces*

* 0·357 in. dia. × 2 in., parallel with weld at the centre.

2 *Surface view of heat-affected zone cracking, showing crack following contour of weld run* Actual size

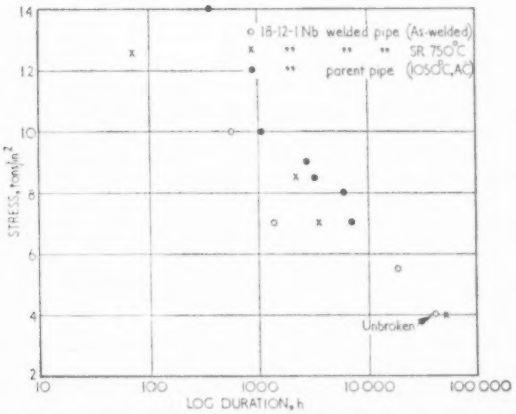
3 *Heat-affected zone cracking produced under laboratory conditions*

were taken from an 18–12–1Nb steel pipe ($1\frac{1}{8}$ in. thick by 8 in. bore and solution treated at 1050°C followed by air cooling) which had been welded with Armex GT electrodes. Details of the pipe and weld compositions are as follows:

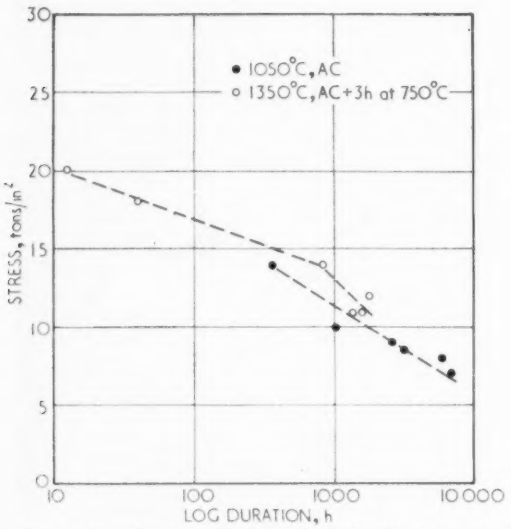
Material	C	Si	Mn	Ni	Cr	Nb	Mo
18–12–1Nb steampipe	0·12	0·37	1·21	11·65	18·54	1·30	...
Armex GT weld metal (typical)	0·08	0·64	2·21	9·18	17·22	...	1·84

After machining, and before creep rupture testing, all test pieces were lightly etched and examined microscopically; no unsound test pieces were used in the investigation.

Figure 4 summarizes the results of stress to rupture tests carried out at 650°C on welded and unwelded test pieces. The welded test pieces were tested both 'as welded' and after stress relieving at 750°C for 3 h. All the welded test pieces failed in the parent metal very close to the weld junction, as in the case of the



4 *Stress to rupture tests at 650°C on welded and unwelded test pieces*



5 *Comparison of rupture behaviour of 18–12–1Nb steam pipe at 650°C (after solution treatments of 1050° and 1350°C)*

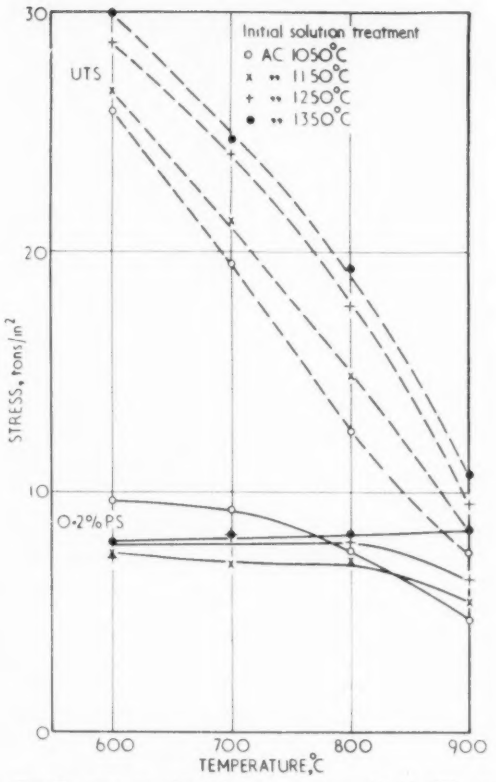
service failures, except for one test (4 tons/in²) lasting about 50000 h, which failed in the weld deposit. The laboratory test pieces differed from the service failures, however, in that cracking was roughly parallel to the weld junction through the section of the test piece. Similar tests on other welded pipes, plates, etc. tested under the same creep rupture conditions also cracked at the weld junctions. However, test pieces stress relieved at 850°C and 1050°C did not fracture at the weld junction, but failed in the parent metal well away from the heat-affected zone.

All the fractures were intercrystalline in character, but it was not possible to obtain actual ductility measurements in the junction zone where cracking was initiated. The impression gained, however, was that the junction zone was greatly lacking in ductility, and it was on this account that it was decided to investigate the strength and ductility of 18–12–1Nb steel after various heat-treatments and at various temperatures.

As failure in most of the creep rupture tests across welded 18–12–1Nb steel occurred in the parent metal close to the weld deposit, it was obvious that any investigations into the behaviour of this steel during welding would have to be based on high solution treatments at temperatures practically up to melting temperature. This temperature range was determined by micro-examination of solution treated samples which indicated that the boundary fusion on a micro-scale occurred just above 1350°C. The top solution treatment temperature used for subsequent investigations into ductility and strength aspects was therefore restricted to 1350°C maximum.

Creep rupture tests on 18–12–1Nb parent material after high solution treatments

Since practically all the creep rupture tests across welded 18–12–1Nb components failed in the weld heat-affected zone when tested 'as welded' or after stress relieving at 750°C, it appeared that the parent material in the vicinity of the weld junction had a lower rupture strength than either the weld deposit or the unaffected

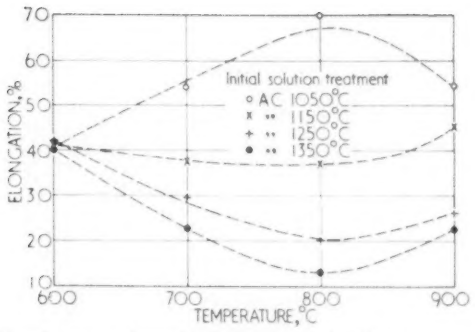


6 *Tensile tests at elevated temperatures, 18–12–1Nb steel specimens 0·357 in. dia.*

parent material. Considering heat-treatment effects alone, this behaviour was somewhat surprising since it was known that the creep and rupture strength of 18–12–1Nb steel tended to improve with increasing solution treatment temperature. This is confirmed in Fig.5, which compares the rupture behaviour of pipe material in the 1050°C AC condition with that of material resolution treated at 1350°C, and given a simulated stress relieving at 750°C. While the 1350°C+ 750°C treatment only marginally improves the rupture strength in this instance, it is obvious from the results that high solution treatment alone is not the controlling factor in regard to the apparent loss of strength near to the weld junction.

A further feature was considered in this respect; there is evidence that the contraction stresses in heavily restrained weldments due to the cooling from the welding temperatures actually cause a 'warm working' effect in both the weld metal deposit and the weld heat-affected zone of the parent material. 'Warm working' of austenitic steels is well known in regard to its effect of improving creep and rupture strength[13]; it is probably less well known that either cold working or warm working austenitic steels reduces rupture ductility.

Rupture tests at present in progress on Armex GT weld metal deposited under conditions of high and low restraint indicate that rupture behaviour is significantly improved for weld metal deposited under highly restrained conditions, such as occur during fabrication of butt welds in thick-walled piping. The general



7 *Tensile tests at elevated temperatures, as in Fig.6*

behaviour is consistent with the weld deposit undergoing 'warm working' by the plastic strains occurring during cooling from weld heat. It seems reasonable to expect that the parent metal adjacent to a weld will suffer plastic deformation in a similar manner to the actual weld deposit, yet on the evidence of the rupture testing, the rupture strength of the weld junction is lower than would be expected from its treatment and subsequent 'warm working'. From this, it would appear likely that failures near the weld junction in creep rupture specimens cut from 18–12–1Nb weldments might be associated more with very poor creep ductility of the junction material than lack of rupture strength *per se*. It follows from this that the cracking troubles experienced during fabrication or service could be due to a similar cause.

As a result of this, the hot ductility of 18–12–1Nb steel was examined more closely, and a preliminary series of tensile tests was carried out at various temperatures, after initial solution treatments up to 1350°C.

Tensile tests at elevated temperatures

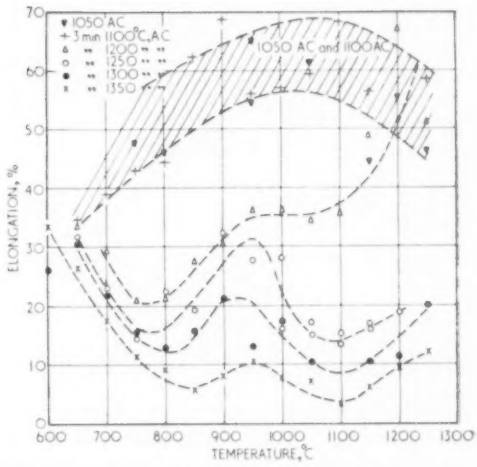
The tests were carried out on bar material with the following composition:

C	Si	Mn	Cr	Ni	Nb
0·09	0·73	1·37	17·28	11·75	0·70

Initial solution treatment temperatures of 1050, 1150, 1250, and 1350°C were used, and tests carried out at 600, 700, 800, and 900°C for each solution treatment. Figure 6 shows ultimate strength and 0·2% proof stress plotted against test temperature, and Fig.7 shows tensile ductility plotted against test temperature. Increasing solution treatments gave a slight increase in ultimate strength, with a slight lowering of 0·2% proof stress, but the proof stress levels after the higher solution treatments were maintained to higher test temperatures. Thus, at 900°C, the 0·2% proof stress, after an initial solution treatment of 1350°C, was as high as the ultimate stress of the 1050°C treated samples.

From Fig.7, it will be seen that increasing solution treatment temperatures resulted in a marked deterioration of ductility in certain temperature ranges.

These results were obtained by normal tensile testing of 0·357-in. dia. specimens (2 in. gauge length). Test pieces were heated up to test temperature in 30–45 min, but soaking time at test temperature was somewhat shorter than normally used and approximately 15 min was found to be sufficient to obtain reasonably steady temperature conditions.



8 *Elongation values obtained after various solution treatments*

A loading rate of 5–6 tons/in²/min was used up to the yield point, and the testing time was 5–10 min to fracture after straining started.

An extensive series of tests was programmed based on this preliminary work, and a simpler method of testing was adopted, with the aim of investigating ductility effects only.

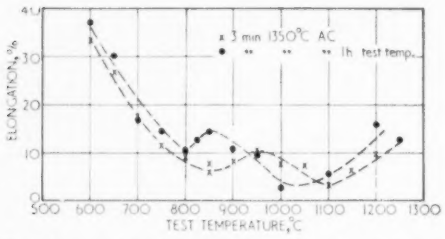
A ¾-ton Denison rupture test machine was modified to carry out short-time tensile tests. Test pieces 0·1795 in. dia., with a ½ in. gauge length, were assembled in gluts and inserted into a warm furnace, the time to reach test temperature averaging 15–20 min. To avoid undue soaking times at test temperatures, for the main series of tests, straining started when the temperature recorded by the thermocouple attached to the specimen was within 2–3°C of the required temperature. Straining was carried out manually and time to fracture ranged from 4–8 min (this gave a strain rate of about 0·03 in./in./min up to yielding). With this equipment it was possible to cover test temperatures from 600° to 1250°C.

To ensure consistency in heat-treatment, all test blanks were rough turned to ½ in. dia. before treatment and held at treatment temperature for about 3 min at solution temperatures of 1250°C and above, and for 15 min at solution treatment temperatures below 1250°C.

Several casts of 18–12–1Nb steel were examined and all showed similar trends in ductility to the cast reported in Fig.7, so that further testing was confined to the first cast.

In the new series, test temperatures were extended to 1250°C and, as mentioned the tests were carried out on the small Denison machine using the short soaking time procedure.

The elongation values obtained on this material after various initial solution treatments are given in Fig.8, the values resulting from the AC 1050°C and AC 1100°C solution treatments comparing reasonably well with normal tensile tests reported on 18–12–1Nb material. A drop in ductility at test temperatures in excess of 1200–1250°C was anticipated from the results reported by Puzak and Rischall.[12]



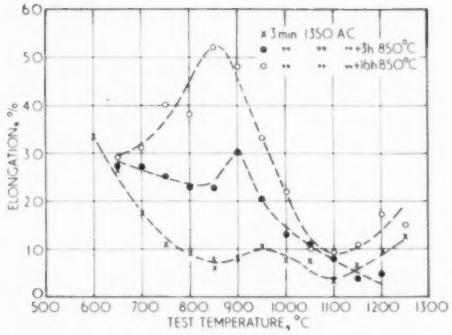
9 *Effect of pre-soaking for 1 h at test temperature before straining*

Raising the initial solution treatment temperature to 1200°C resulted in the occurrence of a ductility trough at 750–800°C, followed by recovery. After still higher initial solution treatment temperatures (1250–1350°C), the trough initially at 750–800°C was displaced to slightly higher temperatures, finally giving a minimum in ductility at 850°C for an initial solution treatment of 1350°C. In addition, recovery from this ductility trough occurred only up to temperatures of around 900–950°C and was followed by a further trough at around 1100°C. This second ductility trough resulted in somewhat lower ductility values than did the earlier trough, and values as low as 3·5% elongation at 1100°C were recorded following an initial solution treatment of 1350°C.

Owing to the particular method of testing used, it was difficult to investigate the effects of different straining rates in detail, but the general trend of results obtained indicated that slower straining rates moved the first ductility trough (800–850°C) to a slightly lower temperature. It was considered that longer soaking times at test temperature before straining would probably result in a similar effect, and a further series of tests, after an initial treatment of 1350°C, was carried out, with a soaking time of 1 h at test temperature, instead of straining as soon as the test temperature was approached. Figure 9 shows the results of these tests, compared with those obtained by the original procedure, and indicates that both minima (800–850°C and 1100°C) are displaced to lower temperatures by increasing the time at test temperature before straining.

This latter series of tests also indicated that below 800°C, soaking times of up to 1 h had no significant effect on the ductility values, but that a significant increase was obtained at 850°C. A further series of tests was therefore carried out, again after an initial solution treatment of 1350°C, but with *every* test treated for 3 h at 850°C before testing in the normal manner, i.e. with a very short soaking time near the test temperature before straining. Results from this series are shown in Fig.10 with the original 1350°C solution treated results for comparison.

A marked reduction of the first ductility trough was observed as a result of this pretreatment at 850°C, and the specimen tested at 850°C after 3 h soaking at 850°C gave an elongation at rupture of 23% compared with 14% after 1 h presoak and 6% with no presoaking before testing. A further series with a pretreatment of 16 h at 850°C after the initial AC 1350°C solution treatment, resulted in even further improvement, and the original ductility trough at 800–850°C was virtually eliminated by this treatment. These results are also plotted in Fig.10. At test temperatures above



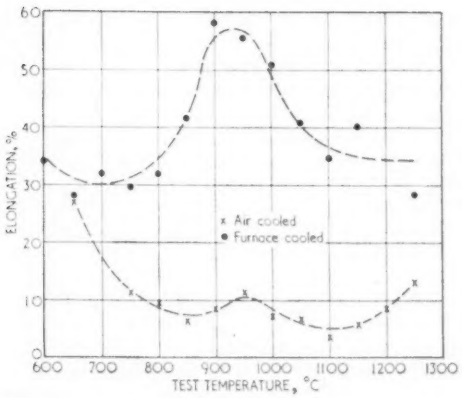
10 *Effect of pre-treatment at 850°C before straining*

900°C, however, the elongation values after both the 3 h and 16 h pretreatment at 850°C dropped rapidly and were of the same order as those obtained from the straight 1350°C AC treatment.

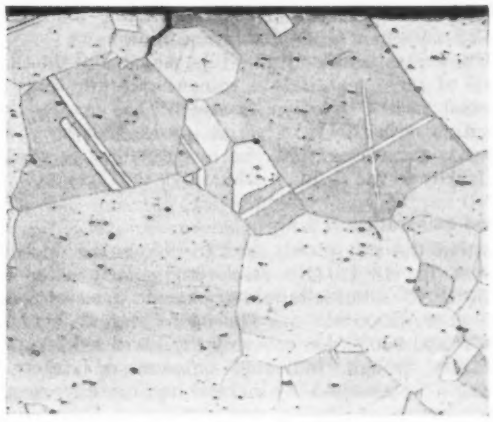
These results demonstrated that a suitable preageing treatment could eliminate the ductility trough experienced during testing up to 900°C, and it was considered that a slower cooling rate from the initial solution treatment temperature should show similar effects. Further samples were therefore treated at 1350°C for 3 min, followed by cooling in the furnace instead of air cooling as previously. This resulted in a cooling rate of about 400°C/h at 1100°C and 220°C/h at 850°C. Tensile tests between 600° and 1250°C resulted in the ductility values shown in Fig.11, with the original values for 1350°C AC initial treatment for comparison. These results indicated that the two low ductility troughs were practically eliminated by this treatment, though two troughs at higher levels of ductility were obtained.

The ductility curves shown in Fig.8 indicated that with the 1350°C solution treatment, elongation values as low as 6% were obtained at 850°C. A series of tests was strained various amounts at 850°C without continuing to fracture, and subsequent micro-examination indicated that surface intergranular cracking occurred after strains as low as 1%. This feature is shown in Fig.12. Similar tests at 650°C and 900°C, where the ductility values at rupture were higher, indicated that proportionally higher strains were necessary to produce similar intergranular surface cracking. This feature was not found, however, at 1100°C, i.e. at the temperature of the second ductility trough. Here, tests on material initially treated at 1350°C AC showed no intergranular surface cracking after strains as high as 2%, even though the ductility at rupture at this temperature was as low as 3·5% elongation, and the mode of fracture was intergranular.

There appeared to be several interesting features connected with the cracking found at low strains at 850°C. Firstly, although at 850°C surface intergranular cracking was found with strains of about 1%, these remained only about 1 grain deep after strains as high as 3 to 4%, although more numerous. A further series of samples strained approximately 3% at 850°C and containing surface cracks was then broken at temperatures as low as 600°C and showed reasonable ductility at rupture (12% at 600°C) which suggested that between 600° and 850°C these intergranular



11 *Comparison of air cooling and furnace cooling from 1350°C*



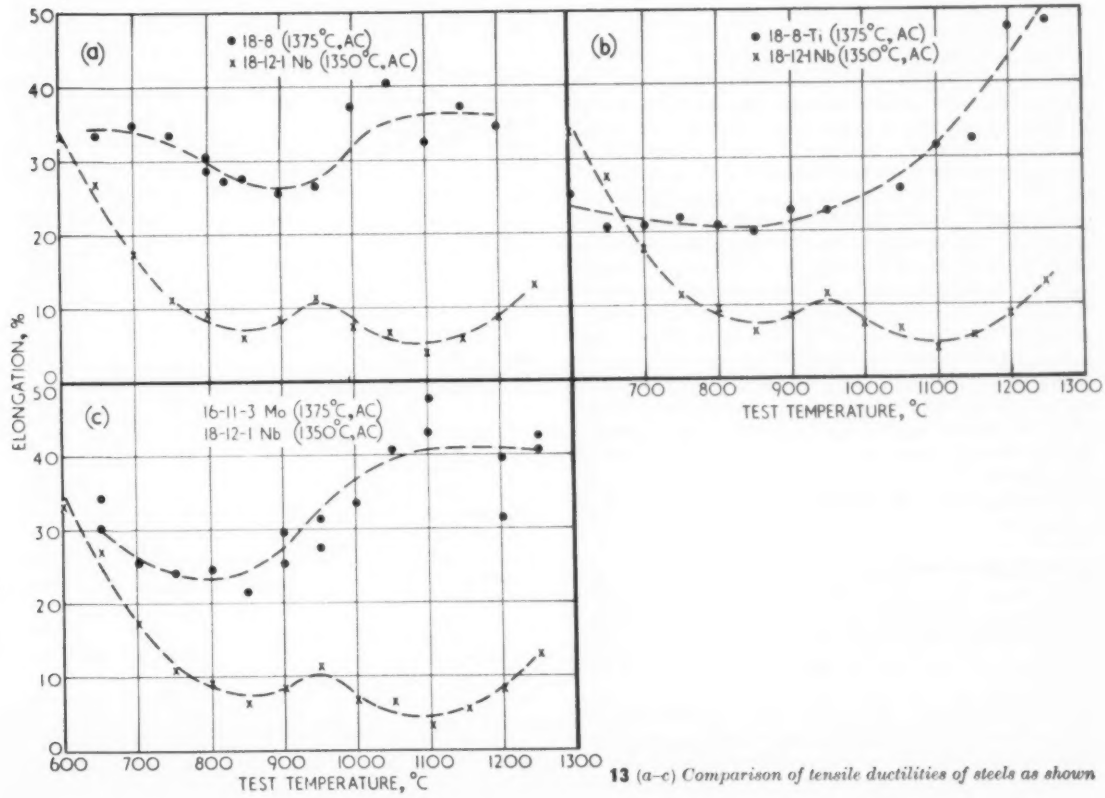
12 *Surface cracks on tensile specimen strained 1% at 850°C after initial solution treatment of 1350°C, air cooled* ×100

cracks, although readily initiated, did not readily propagate under the particular test conditions employed.

Although the object of this investigation was to study ductility behaviour of 18–12–1Nb steel, it was considered useful to investigate other austenitic steels under similar test conditions. For this purpose, a limited amount of testing has been carried out on a non-stabilized 18–8 steel, titanium-stabilized 18–8–Ti and 17–11–Mo steels. The results of these tests follow:

	C	Si	Mn	Cr	Ni	Ti	Mo
18–8	0·07	0·54	1·84	17·80	8·70	...	...
18–8–Ti	0·08	0·88	1·64	18·03	8·43	0·65	...
17–11–Mo	0·05	0·19	1·69	16·38	10·50	...	2·55

Preliminary examination indicated that these steels could be treated up to 1375°C without causing boundary fusion effects, and this was adopted as the initial solution treatment temperature for this series. Ductility curves are shown in Fig.13 for this series



13 (*a–c*) *Comparison of tensile ductilities of steels as shown*

of steels, comparisons in each case being made with 18–12–1Nb steel (1350°C solution treatment).

These results show that while the general ductility level of these materials is somewhat lower than for normal 1050°C solution treatment, the marked low ductility troughs found with 18–12–1Nb steel are absent, suggesting that the low ductility is related primarily to the niobium addition.

Micro-examination

Most of the test pieces used to obtain the ductility curves for 18–12–1Nb steel, particularly after the 1350°C AC initial solution treatment, were sectioned for micro-examination. Specimens tested between 600°C and 950°C, i.e. covering the lower temperature ductility trough, indicated only minor structural changes as assessed by normal optical microscope. Boundary precipitation was obvious on these specimens, and assumed to be $M_{23}C_6$ type carbide, from previous experience with this steel. Specimens tested at 800–900°C also showed some change in the appearance of the matrix, although actual precipitation could not be resolved.

In the range 950–1050°C, boundary precipitation was apparently less than at lower temperatures, but above 1050°C precipitation on boundaries and in the grain became increasingly easier to resolve. At the highest temperature examined, i.e. 1250°C, the amount of precipitation appeared to be decreasing generally.

It became obvious that the effects responsible for the wide variation in tensile ductility could not be resolved by optical microscope examination, and for more detailed study selected samples were examined by electron microscopy by Associated Electrical Industries (Manchester)Limited. Their detailed examination will form the subject of a separate paper, but preliminary indications are that the ductility trough at 850°C is associated with a very finely dispersed precipitation of niobium carbide. No such precipitation is apparent at 650°C, where ductility at rupture was relatively high. At temperatures above 850°C, the precipitates in the matrix appear to be in an over-aged condition, the particle size and dispersion increasing with temperature. Type $M_{23}C_6$ carbide was detected on grain boundaries in samples tested up to at least 1050°C. A marked denudation effect in regions adjacent to the boundaries was also found with samples tested at 950–1100°C.

Pre-ageing at 850°C was found almost to eliminate the lower ductility trough in subsequent testing, and it was confirmed that this treatment caused precipitation of discrete particles of niobium carbide generally throughout the matrix, thus significantly reducing subsequent strain-induced precipitation effects during testing.

DISCUSSION OF RESULTS

It has been shown that after high solution treatments comparable with those attained in weld heat-affected zones, 18–12–1Nb steel exhibits very low tensile ductility in two temperature ranges, one around 850°C and one around 1100°C. This behaviour is related to solution treatment temperature, the ductility decreasing with increasing solution treatment temperature. On the evidence available, both the low ductility troughs appear to be due to carbide solution and precipitation mechanisms, and, so far, are peculiar to niobium-stabilized steels. A super-saturation effect caused by rapid cooling from high solution treatment temperatures is necessary to promote these effects.

The ductility trough found at 850°C is due to strain-induced precipitation of niobium carbide during actual testing. This apparently results in strengthening of the matrix compared with boundaries, and causes most of the straining to occur in boundary regions with resultant intergranular failure at low overall elongations. Boundary precipitation of $M_{23}C_6$ type carbide may also have contributed to some degree. The strain-induced precipitation occurs rapidly at temperatures approaching 850°C, as these effects were found in tests of only a few minutes' duration. Pre-ageing at 850°C has been shown virtually to eliminate subsequent strain-induced precipitation, as assessed by this particular method of testing.

The second ductility trough at 1100°C is obviously also dependent on solution and precipitation effects, as judged by the various ductility trends reported, and so far no evidence has been found to show that it is associated with precipitation of any phase other than niobium carbide.

Rapid over-ageing of matrix precipitates at test temperatures above 850°C causes some recovery of ductility, and though the evidence indicates that particle size and dispersion increase with increasing test temperature, recovery of ductility occurs only up to 950°C and is followed by the second ductility trough at 1100°C.

Supplementary tests to investigate this higher temperature ductility trough indicate that the mechanism causing these low ductility values is operative only in this high temperature range. Tests 'pre-conditioned' to give the structures found in the tensile tests strained at 1000–1100°C were found to give good ductilities at room temperature, 600°C, 800°C, and 850°C. The 'pre-conditioning' was carried out on tests initially solution treated at 1350°C AC, which were then re-heated to temperatures of 1000°C and 1100°C, held for times of 1 h and 5 min respectively, then rapidly cooled.

Other tests were strained up to 2% at 1100°C without testing to failure, cooled rapidly, then re-tested at various temperatures.

The following test results were obtained from this series:

All tests initially 1350°C AC		
Pre-conditioning treatment	Test temp., °C	Ductility, %
1 h 1000°C AC	600	29
5 min 1100°C AC	20	50
5 min 1100°C AC	600	30
5 min 1100°C AC	800	29
5 min 1100°C AC	850	32
5 min 1100°C + 1% strain at 1100°C AC	800	32·5
5 min 1100°C + 1% strain at 1100°C AC	850	30
5 min 1100°C + 2% strain at 1100°C AC	20	50
5 min 1100°C + 2% strain at 1100°C AC	600	25

These results indicate that the structures produced either before or during straining at test temperatures in the 1000–1100°C range do not appear to affect

ductility adversely at any other temperature. In fact, the effects normally found at 850°C, i.e. the lower ductility trough after 1350°C AC are virtually eliminated by the 1100°C pretreatment.

However, pre-ageing at 850°C, while practically eliminating the lower ductility trough, did not significantly alter the ductility behaviour in the 1100°C region. Examination of specimens tested above 950°C showed well dispersed over-aged precipitation in the matrix with marked boundary depleted regions, this latter feature persisting up to 1100°C. It is tentatively suggested that at temperatures as high as 1100°C, the matrix strength is so low that the dispersion of these over-aged precipitates, coupled with boundary depletion effects, could again result in straining occurring mainly in boundary regions during testing and results in low overall elongation with intergranular failure.

At even higher temperatures, 1200–1250°C, which are approaching the initial solution treatment temperature ranges, the amount of precipitation occurring at test temperature becomes progressively less and some ductility recovery should, and does, occur.

Furnace cooling from 1350°C resulted in precipitation of exceptionally coarse particles which were well dispersed, and after re-heating to test temperatures around 1100°C, the structure was not significantly altered. This conforms with the good ductility values recorded after the furnace cooling treatment.

It is interesting that the ductility curves for Type 347 steel, reported by Nippes *et al.*, showed no indication of a ductility trough at around 1100°C with tests carried out during cooling from temperatures of 2400–2500° (1315° to 1370°C). This can probably be explained by the different testing techniques employed compared with the present investigation. The work of Nippes *et al.* involved tensile testing at various temperatures while cooling from the initial high solution temperature, the cooling rate being very rapid, only a few seconds being involved in cooling from 2500° to 2000°F (1370 to 1093°C). The tests described in this present investigation were carried out after cooling to room temperature from the high solution treatment temperature followed by re-heating to test temperature before tensile testing. The difference in precipitate size and dispersion due to these different thermal cycles may well explain the absence of the low ductility effects around 1100°F in the work of Nippes *et al.*

After examination of the low ductility behaviour of 18–12–1Nb steel in relation to weld heat-affected zone cracking, it seems reasonable to assume that the plastic deformation which occurs in heavily restrained welds during cooling from the welding cycle takes place to some degree at temperatures which are associated with the low ductility troughs. Of these two troughs, that found at 1100°C is considered less significant than that found at temperatures of around 850°C because, of the two regions investigated, only that around 850°C appears to show the combined effects of strain and temperature as assessed by the resultant strain-induced precipitation. The effects found around 1100°C appear to be somewhat peculiar to the testing techniques employed, and at this temperature, 'conditioning' of the material by the combined effect of strain and temperature does not appear to be operative. The mechanism involved in producing the low ductility region around 850°C, coupled with the fact that in this region cracking can commence at strains of 1% or less (which is not true of the 1100°C ductility region) indicates that the lower ductility trough at 850°C is more likely to be involved in weld cracking troubles.

Neither of the troughs investigated is indicated in the ductility curves given by Nippes *et al.* from tests carried out at various temperatures during cooling from 2450°F (1345°C).

The low ductility effects observed by Nippes *et al.* at very high temperatures are not considered to have primary significance in regard to weld cracking, except in causing fissures near to the weld junction. These are relatively wide and shallow, and are different in nature from the fine intergranular cracking found in the short-time tensile test pieces in the region of 850°C, and in cracks produced during stress relieving of heavily restrained welds. In the latter instance, the fissures and fine intergranular cracks can often be observed in the same section, and it is apparent that the fine intergranular cracking has propagated from the surface fissures. Another feature which can cause surface fissures is high-temperature corrosion of the surface of the weld heat-affected zone by the welding flux (electrode coating) used. Such attack is usually a shallow intergranular surface penetration and has been produced artificially under laboratory conditions on samples which have not been subjected to an actual welding cycle.

From the results of tests reported in this paper, it is postulated that the parent metal adjacent to a weld junction is plastically deformed during cooling from welding heat, and that sufficient straining occurs in the 850°C region to condition the material so that subsequent ductility is very limited. In this condition, any further small amounts of strain, such as may occur as creep strain during stress relieving or under service conditions, would be sufficient to initiate cracking, or at least to cause rapid crack propagation from surface notches or fissures existing for the reasons given above.

This conditioning of the weld heat-affected zone by straining during cooling from welding is probably also the major reason for the position of the failures which occurred during rupture testing of welded specimens rather than any effect produced by the low ductility regions investigated by Nippes *et al.* When tested as welded, or stress relieved at 750°C, failure occurred at the weld junction, but when tested after stress relieving at 850°C or 1050°C failure occurred in the parent metal well away from the weld heat-affected zone.

Similar effects have been reported by Christoffel[14] in notched rupture tests on Type 347 pipe material, who found that the welding cycle produced a drastic reduction in notched rupture strength of weld heat-affected zone material which was not produced by a weld thermal cycle alone (without plastic straining).

Further work is clearly required if only to establish a more direct relationship between the 850°C trough and the behaviour of a welded joint.

A better understanding of the 1100°C low ductility region is also needed to assess whether this plays any part in weld cracking troubles.

CONCLUSIONS

It has been shown that 18–12–1Nb steel exhibits very low tensile ductility behaviour at 850°C and 1100°C

after being subjected to high solution treatment temperatures, such as are attained in weld heat-affected zones. The low ductility trough associated with test temperatures around 850°C was found to be caused by strain-induced precipitation of niobium carbide in the austenite matrix, and it is suggested that this mechanism is probably responsible for the susceptibility of 18–12–1Nb weldments to weld junction cracking. The second ductility trough at 1100°C is not considered to have an important bearing on this particular problem.

ACKNOWLEDGMENTS

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Heat-affected zone cracking in welded high-temperature austenitic steels

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SYNOPSIS

Heat-affected zone cracking during welding and during post-weld heat-treatment has been investigated in 18Cr–12Ni–1Nb, 18Cr–10Ni–3Mo, and 18Cr–3Ni type steels using a simple welding test. Metallographic investigation has been carried out using direct carbon extraction replica techniques and the extracted carbides were examined by electron microscopy and electron diffraction.

The results indicate that limited cracking occurs during welding, followed by extensive intercrystalline cracking during subsequent heat-treatment. The mechanisms of the two forms of cracking are quite different, that during welding being due to tearing at regions of local liquation and that during heat-treatment to loss of ductility due to strain-induced precipitation. 1863

WITH THE GREATER USE in recent years of Cr–Ni austenitic steels for high-temperature applications, the problem of heat-affected zone cracking has become very serious. This type of failure usually occurs in piping with a wall thickness greater than $\frac{3}{4}$ in., and has been reported both during welding and during stress relieving, while a number of failures have apparently occurred during service. The position of the cracks has been usually reported as being in the heat-affected zone approximately parallel to the fusion line of the weld. The cracking has been generally agreed to be intergranular in character. A review of the current literature has recently been made by Borland and Younger.[1] The current theories of cracking appear to fall into two broad groups, those dealing with cracking during welding and those with cracking during heat treatment.

Pellini and his associates[2] have found that cracking may occur during welding and have suggested a mechanism in which cracking occurs by hot tearing due to the liquation of low melting point segregates. Nippes and his associates[3] believe that damage to the heat-affected zone occurs when it is cycled to a high peak temperature. They have obtained experimental evidence to show that loss in ductility occurred at temperatures approaching the solidus. During cooling some materials regained ductility very slowly, and cracking susceptibility was believed to be associated with this type of behaviour. Nippes *et al.*[4,5] also carried out creep-rupture tests at 650°C on materials previously given a heating and cooling cycle, but concluded that the cycle did not impair the mechanical properties of the heat-affected zone during subsequent reheating.

Emerson and Jackson[6] suggested that failure could occur during stress relieving by a creep-rupture mechanism if the applied stress were sufficiently high,

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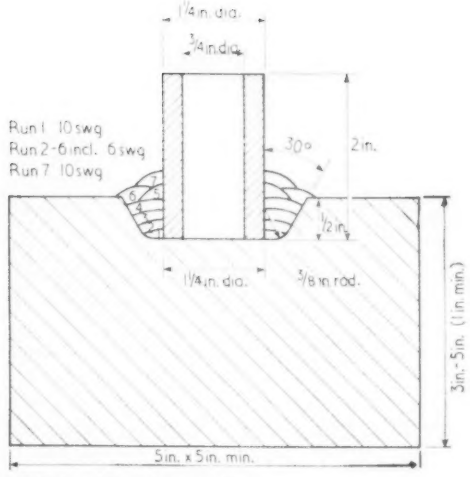


1 *BWRA welding test procedure*

and that the stress level was the predominant factor. Poole[7] also believed that cracking occurred by a similar mechanism.

The present work has been particularly concerned with the investigation of cracking in the heat-affected zone of 18Cr–12Ni–1Nb steel. This steel is used in superheater headers and for the stub pipes which are welded to them. Cracking has been observed both in as-welded joints after a period of service, and also following stress relieving immediately after welding. To study the problem in the laboratory a simple test has been devised in which cracking can be successfully reproduced. From the preliminary work it was concluded that cracks usually propagated during stress relieving from small hot tears originally formed during welding. The welding test has also been used to investigate the cracking susceptibility of other austenitic steels. It was thought that the microstructure developed in heat-affected zones would be a most important factor in determining the cracking susceptibility. Accordingly a microstructural survey of the steels tested has been carried out using the electron microscope. The results of the welding tests and the associated metallographic work are given in this paper followed by a discussion of the mechanism of cracking.

TABLE I Nominal composition of electrodes used

Electrode	C	Mn	Si	Ni	Cr	Mo	Nb	Cu	Co	W
A	0·09	1·2	0·46	9·2	18·8	...	0·7	2·0	...	...
B	0·3	2	0·6	15	17	3	2	...	7	...
C	...	...	...	...	33	...	...	...	55	6

EXPERIMENTAL TECHNIQUE

BWRA cracking test

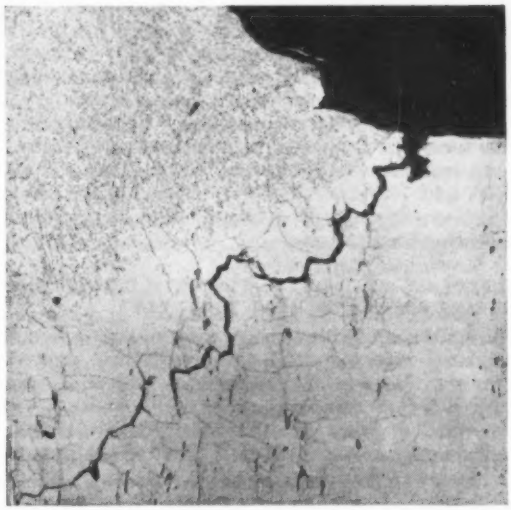
This test, devised by Borland *et al.*[8] consisted of a block of the material to be tested (of minimum dimensions 5 in. × 5 in. × 3 in.) and a tube of the same material with a maximum outside diameter of 1¼ in. The joint geometry and the welding procedure conformed closely to those used in service welds. The metal arc process with the electrodes listed in Table I was used throughout. Figure 1 is a diagram of the test assembly.

After welding, the heat-affected zone was examined for signs of cracking at a magnification of ×20 and metallographic sections were taken through any suspect areas. For heat-treatment tests a similar weld was made and subsequently the test assembly was heated in a muffle furnace at 100°C/h to the required temperature. After heat treatment the assembly was furnace cooled and again examined for cracking.

Metallographic investigation

After heat treatment, pieces were cut from the test assembly so that a transverse section of the heat-affected zone of the base plate could be examined. The metallographic specimens were ground, abraded, electropolished, and etched in a mixture of hydrochloric, nitric, and sulphuric acids. Carbon extraction

2a *Hot tearing in the heat-affected zone of 18Cr–12Ni–1Nb-type steel after welding; etched in oxalic acid* ×200



2b *Cracking in the heat-affected zone of 18Cr–12Ni–1Nb-type steel after 100 h at 750°C; etched in oxalic acid* ×50

TABLE II Effect of time and temperature on heat treatment, cracking of 18–12–1 steel

Temperature, °C	Heat treatment time, h				
	0·1	1	10	100	1400
750	nil	cracked	cracked	cracked	cracked
650	nil	nil	nil	cracked	cracked
550	nil	nil	nil	nil	cracked

replicas were then taken using the technique developed by Smith and Nutting.[10] The replicas were examined in a Siemens Elmiskop 1 electron microscope, the selected area electron diffraction technique being used, where possible, to identify precipitated phases.

OBSERVATIONS OF CRACKING DURING WELDING AND DURING HEAT TREATMENT

Cracking of 18–12–1Nb steel during welding

During welding with electrode *A* (Table I) a small crack was produced in the heat-affected zone of all the joints tested. A typical photomicrograph is shown in Fig.2*a*. Cracks of this kind appeared to be intergranular and were rarely more than 0·03 in. in depth. They occurred very close to the fusion line and their general direction was perpendicular to the surface.

It was suggested by Houldcroft[9] that by analogy with non-ferrous systems, cracking of this type might be eliminated by using an electrode with a lower solidus temperature. This was also in accordance with the ideas of Pellini *et al.*[2] who had postulated a hot-tearing mechanism to explain cracking found during welding of similar steels. Accordingly welds were made with a hard-facing electrode (electrode *C*, Table I) whose solidus temperature had been found to be 50–100°C less than that of electrode *A*. No cracks were formed during welding with the low-solidus electrode.

Cracking of 18–12–1Nb steel during heat treatment after welding

On reheating welded 18–12–1Nb steel to 750°C for over 1 h, extensive intercrystalline cracks propagated from the root of the hot tear formed during welding. A typical photomicrograph is shown in Fig.2*b*. The cracks in this case were confined to the region of the heat-affected zone and were of considerable extent sometimes reaching depths of $\frac{1}{2}$ in. or more.

The onset of this form of cracking depended on both the time and temperature of reheating. Results of tests carried out at reheating temperatures of 550°, 650°, and 750°C are shown in Table II. The time at which cracking began decreased with an increase in reheating temperature.

EFFECT OF WELD METAL STRENGTH ON CRACKING DURING POST-WELD HEAT TREATMENT

During the course of testing it was observed that the degree of cracking which occurred during post-weld heat treatment in 18Cr–12Ni–1Nb type steel varied with differing compositions of weld metal. It was suspected that the variations bore some relationship to the hot strength of the weld metal. For this reason the test was graded by welding with three different weld metals before heat treatment. The compositions of the three electrodes used are given in Table I, the hot strength at 750°C increasing in the order $A < B < C$.

TABLE III Analysis of materials tested

Material type	Composition C	S	P	Si	Mn	Cr	Ni	Nb	Mo
18–10–Mo	0·04	0·016	0·018	0·22	1·59	16·56	10·54	...	2·25
18–8	0·07	0·025	0·025	0·60	0·90	18·78	8·18	...	...
18–12–Nb	0·085	0·005	0·029	0·64	1·48	17·27	12·20	0·87	...

Tests were carried out on 18Cr–8Ni and 18Cr–10Ni–3Mo type steels as well as on 18Cr–12Ni–1Nb type steel. In almost all cases (electrode *C* welds excepted) a hot tear occurred during welding but a record of cracking was only made if the degree of cracking increased markedly during heat treatment. This distinction was usually extremely clear.

Where cracking occurred at all during heat treatment the degree of cracking increased with increasing strength of weld metal. The same pattern of behaviour was observed in niobium-bearing and unstabilized steels. No cracking was observed in the case of the steel stabilized with Mo alone even when welded with electrode *C*. The analyses of the materials tested are shown in Table III and the results of testing in Table IV. The heat treatment was in all cases 100 h at 750°C.

METALLOGRAPHY OF THE HEAT-AFFECTED ZONE OF WELDED SPECIMENS

These results were all obtained from extraction replicas taken from a region within three grain diameters of the fusion line.

18Cr–12Ni–1Nb steel welded with electrode A and heat treated 100 h at 750°C (cracked during heat treatment)

At low magnification in the light microscope it was observed that heavy precipitation had occurred at the grain boundaries of this specimen but the electromicrograph (Fig.3*a*) showed that in addition a precipitate of numerous fine particles was present within the grain. A traverse across the heat-affected zone showed that the precipitated particles were most numerous within a few grains of the fusion line. A micrograph of the parent plate outside the heat-affected zone is shown in Fig.3*b*. Selected area electron diffraction studies of the fine precipitate showed that it was NbC. The morphology of the particles precipitated at the grain boundaries varied widely throughout the specimen but Fig.3*a* is fairly representative of the heat-affected zone. Where electron diffraction patterns could be obtained from grain boundary carbides they showed $M_{23}C_6$-type structure.

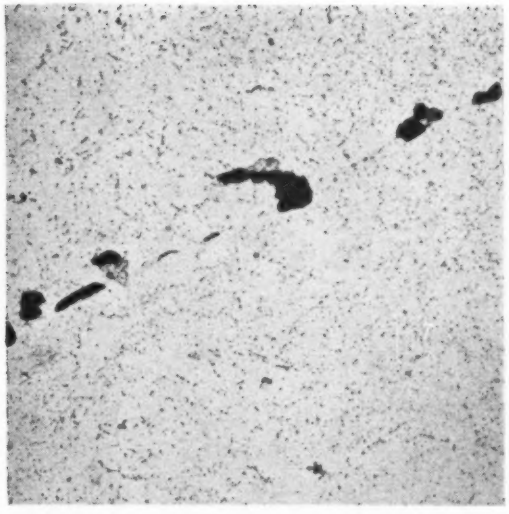
18Cr–10Ni–Mo steel welded with electrode A and heat treated 100 h at 750°C (no cracking during heat treatment)

This specimen again showed heavy grain boundary precipitation at low magnification but electron micrographs (e.g. Fig.3*c*) showed a marked difference from

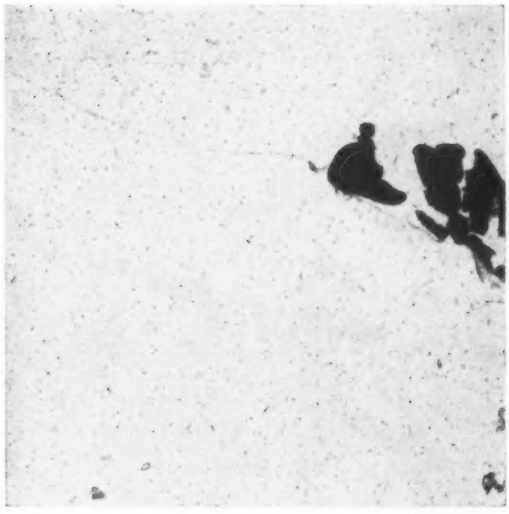
TABLE IV Results of base metal cracking tests after 100 h at 750°C

Electrode type	Material type 18–10–Mo	18–8	18–12–Nb	18–12–Nb ground out*
A	nil	nil	cracked	nil
B	...	cracked	cracked	...
C	nil	cracked	cracked	cracked

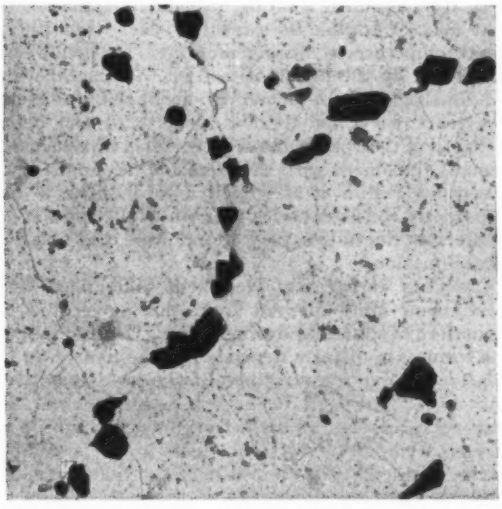
* Fusion line ground out with tungsten carbide tool before heat treatment



3a *Heat-affected zone of 18Cr–12Ni–1Nb-type steel after 100 h at 750°C, welded with electrode A, carbon extraction replica* ×20000



3c *Heat-affected zone of 18Cr–10Ni–3Mo-type steel after 100 h at 750°C, welded with electrode A, carbon extraction replica* ×20000



3b *Parent plate of 18Cr–12Ni–1Nb-type steel after 100 h at 750°C, carbon extraction replica* ×20000

the 18Cr–12Ni–1Nb steel. In the Mo-stabilized steel little or no fine precipitate was observed within the grain. This difference was extremely marked whereas no significant differences in the morphology of $M_{23}C_6$ precipitates at the grain boundary were observed.

18Cr–8Ni steel welded with electrode A and 18Cr–8Ni steel welded with electrode C; both heat treated 100 h at 750°C

These specimens were selected in an attempt to determine first whether an intragranular precipitate existed in an unstabilized steel, and second to determine the effect of variations in weld metal strength on the heat-affected zone microstructure. Only the specimen welded with electrode *C* cracked during heat treatment. Electron micrographs of the heat affected zone of the two specimens are shown in Figs.4*a* and 4*b*. In the first specimen there was some precipitation within the grains but it can be seen in the second specimen that the effect of welding with electrode *C* is to cause a marked increase in the number of precipitated particles. Selected area diffraction studies of this precipitate indicated that it was $M_{23}C_6$.

METALLOGRAPHY OF SIMULATED HEAT-AFFECTED ZONE MICROSTRUCTURES

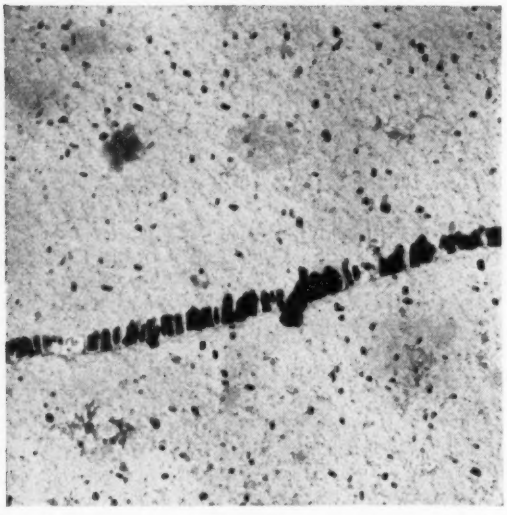
It was considered important at this stage to determine whether the microstructure of the heat-affected zone could be produced by heat treatment alone. For this purpose apparatus was constructed which would subject small metallographic specimens to a thermal cycle similar to that experienced by the heat-affected zone near the fusion zone of the weld. High-frequency heating and argon quenching were used to give heating and cooling rates which corresponded very closely to those measured during welding a $1\frac{1}{2}$-in. thick stainless-steel plate.[4] Specimens heat-treated in this way were then aged.

18Cr–12Ni–1Nb steel thermally cycled, and heat treated 100 h at 750°C

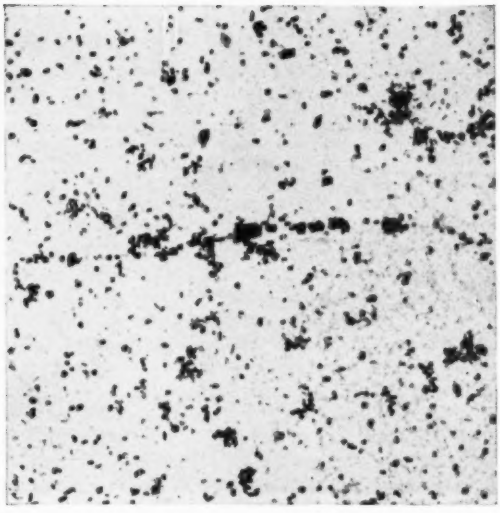
Precipitation of $M_{23}C_6$ occurred at the grain boundaries of this specimen (Fig.5*a*) similar to that found in the heat-affected zone of the weld (Fig.3*a*). However, the nature of the intragranular precipitate was markedly different in the two cases. After thermally cycling, the precipitation sites were considerably rarer and the size of the precipitated particles correspondingly larger than in the heat-affected zone of the weld.

18Cr–12Ni–1Nb steel thermally cycled, plastically deformed, and heat treated at 750°C

It is now well established that precipitation in austenitic steels can be profoundly affected by plastic deformation. Owing to the high thermal contraction stresses occurring during welding it is very likely that



4a *Heat-affected zone of 18Cr–8Ni-type steel after 100 h at 750°C, welded with electrode A, carbon extraction replica* ×20000



4b *Heat-affected zone of 18Cr–8Ni-type steel after 100 h at 750°C, welded with electrode C, carbon extraction replica* ×20000

plastic deformation will occur in the heat-affected zone. For this reason a further series of specimens was subjected to the same thermal cycle and then plastically deformed by bending a segment of the metallographic specimen. This treatment produced an elongation of about 10% in the final polished surface of the specimen.

The combined treatment of deformation and ageing at 750°C produced a fine dispersed precipitate which is similar to that found in the heat-affected zone (Fig.3*a*). Continued ageing at 750°C caused an increase in the number of precipitated particles. Figure 5*b* shows the microstructures after 100 h. The precipitate would thus appear to be strain induced and the much larger deformation of the metallographic specimen than of the actual weld explains the finer dispersion shown in Fig.5*b*. Selected area diffraction from the grain interior revealed a strong niobium carbide (NbC) diffraction pattern.

18Cr–8Ni and 18Cr–10Ni–3Mo steels thermally cycled, plastically deformed, and heat treated at 750°C

The pattern of behaviour in 18Cr–8Ni steel was very similar to that of the Nb-stabilized steel. There was a fine precipitate of $M_{23}C_6$ within the grain while larger $M_{23}C_6$ particles had formed on coarse slip lines. The microstructure was similar to that observed in Fig.4*b*.

In the Mo-stabilized steel little or no intragranular precipitation was observed in the deformed and aged specimens after treatment for up to 100 h at 750°C. The microstructure showed no apparent differences from that found in the heat-affected zone of a weld after a similar ageing treatment (*see* Fig.3*c*).

DISCUSSION

Initiation and propagation of cracking

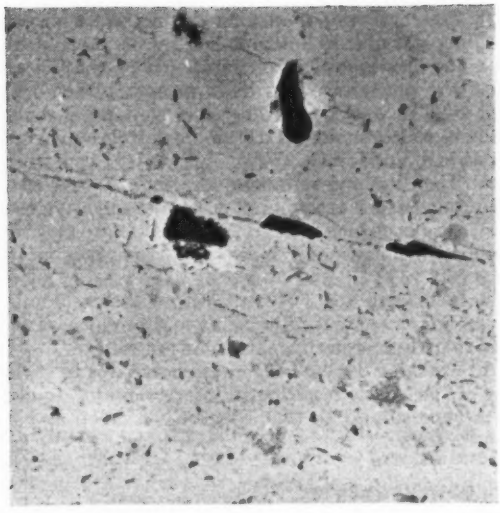
In this investigation intergranular heat-affected zone cracking has only been observed to propagate extensively during heat treatment after welding, but when cracking occurs it is initiated at the root of small hot tears formed during welding. It is thought that the hot tear may be formed as a result of local liquation at high temperatures during welding. This view was supported by the fact that using a weld metal which solidified after the liquefied regions of the base metal, hot tears were not formed in spite of the higher strength of that particular weld metal. This result is in general agreement with the liquation mechanism proposed by Pellini *et al.*[2] Such a mechanism is also apparently consistent with the results of Nippes *et al.*[3] if the low ductilities they found were the result of local liquation.

Cracking during heat treatment takes place by a mechanism quite distinct from that governing the formation of a hot tear during welding. During welding the formation of tears depends predominantly on the relative solidus temperatures of the weld metal and base metal. However, during heat treatment, cracking is increased by using a high-strength weld metal. Thus in the case of electrode *C*, hot tears in the base metal are not formed at all, whereas heat-treatment cracking is much more extensive than with either of the other two weld metals. Furthermore in the molybdenum steel, although hot tears are formed, no further cracking occurs during subsequent heat treatment.

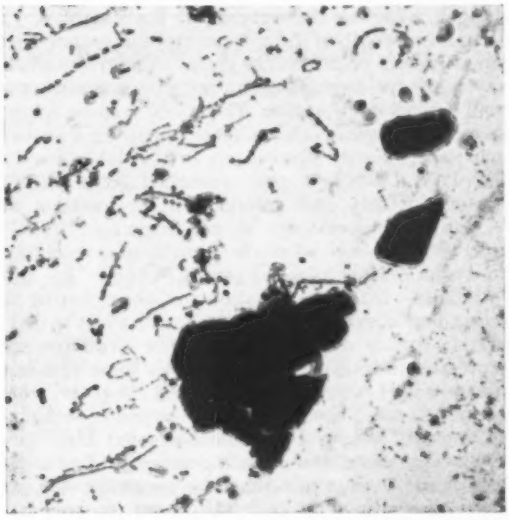
When a hot tear does form in a susceptible steel it can act as a stress raiser, giving a point of easy initiation for cracking during subsequent heat treatment. However, a hot tear is not essential for the development of this form of cracking. The results given in the last column of Table IV show that by grinding out the fusion line of a weld to remove all traces of hot tearing, cracking during heat treatment can be avoided in the heat-affected zone of welds made with electrode *A*, but, under the more severe conditions provided by the use of electrode *C*, cracking during heat treatment is not prevented.

Effect of steel composition

From the results obtained on the three materials so far studied, it is evident that a precipitate of numerous



5a *18Cr–12Ni–1Nb-type steel specimen, thermally cycled and heated 100 h at 750°C, carbon extraction replica* ×20000

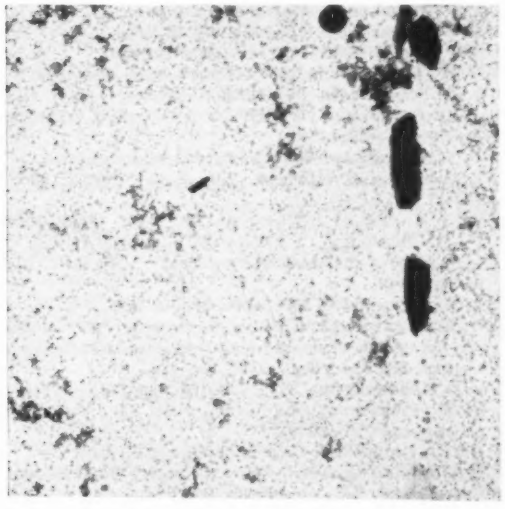


5c *Heat-affected zone of 18Cr–12Ni–1Nb-type steel after 100 h at 750°C, welded with electrode* A, *carbon extraction replica* ×65000

fine particles within the grains is always present after cracking has occurred during heat treatment. In the steel containing molybdenum, which did not crack, little or no intragranular precipitation occurred. In all the steels there was precipitation of relatively large carbides at the austenite boundaries even before heat treatment and, when it was possible to identify these intergranular particles by electron diffraction, they were always found to have the $M_{23}C_6$ structure. In the niobium-stabilized steel the intragranular precipitation was found to be NbC and in the unstabilized steel $M_{23}C_6$.

It is not at present clear why, under the conditions so far studied, the presence of molybdenum inhibits the formation of a fine intragranular precipitate. It should be noted that the carbon level in the Mo steel (0·04%) is appreciably lower than in either of the other two steels. However, it seems unlikely to the present authors that this fact alone could completely explain the different behaviour of the Mo steel. Molybdenum has a higher affinity for carbon than chromium and is known to cause secondary hardening in ferritic steels owing to the formation of a fine intragranular precipitate of Mo_2C.[11–13] However, even in these steels the Mo_2C is frequently replaced by complex molybdenum-based carbides of the $M_{23}C_6$- or M_6C-type after more severe tempering treatments.

In the present case it may be that neither the lattice nor the dislocations present are adequate sites for precipitation of molybdenum carbide at such a low carbon level. The activity of carbon would be expected to be reduced by the presence of molybdenum in solid solution and this might explain why no chromium carbide was precipitated within the grains. On continued ageing it might be expected that the molybdenum and carbon would precipitate in the form of large complex carbides at the boundaries, where $M_{23}C_6$-type carbides are already in existence. To clarify the position it is hoped to investigate Mo-bearing steels at higher carbon levels.



5b *18Cr–12Ni–1Nb-type steel specimen, thermally cycled, plastically deformed, and heated 100 h at 750°C, carbon extraction replica* ×20000

Effect of plastic strain on precipitation

Strain-induced precipitation can occur in austenitic steels if they are either plastically deformed before being aged or alternatively warm worked, dislocations being introduced which can act as preferred sites for precipitation. Convincing evidence that precipitation does occur on dislocations in austenitic steel has been produced by Schrader,[14] Arrowsmith,[15] Pickering,[16] and in the present investigation (*see* Fig.5*c*). As would be expected, increased deformation, in providing more precipitation sites, results in finer and more numerous particles.

The formation of a fine intragranular precipitate stiffens the grain, thus increasing the overall creep

strength, often at the expense of the ductility. The increased strength of the grain causes stress concentration at certain points on the grain boundaries and this may result in intercrystalline failure after only very small overall deformation. The effect may be demonstrated by carrying out hot tensile tests. In a series of such tests recently carried out in this laboratory on 18Cr–12Ni–1Nb-type steel quenched from 1350°C, loss of ductility and intercrystalline fracture was evident at temperatures in the region of 750°C.[17] Kirkby[18] has obtained similar results in a recent and more comprehensive investigation.

At 650°C, little precipitation can occur during the testing period even with plastic strain, while at 900°C precipitation is complete at the few available sites before the testing begins. Thus both these test temperatures give a high ductility result. However, where precipitation can occur during the course of straining, the ductility is low. In the heat-affected zone, however, where from the metallographic evidence it is clear that plastic strain must be occurring, the processes of strain and precipitation are concomitant. Thus loss of ductility, in this case, would be expected up to much higher reheating temperatures than would be indicated by a hot tensile test, where no stress is applied until the test temperature is reached.

By increasing the strength of the weld metal, the total strain taking place in the heat-affected zone is increased and the consequent effect on the microstructure is to make the precipitated particles finer and more numerous. The severity of cracking is, therefore, also increased and even the unstabilized steel cracks if the total strain is sufficient.

The mechanism of cracking

The evidence obtained in the present investigation is that extensive cracking occurs by a creep-rupture mechanism during reheating. So far, it is in agreement with the conclusions of Emerson and Jackson[6] and Poole.[7]

Irvine *et al.* have shown that when strain-induced precipitation takes place in austenitic steels there is a resulting loss of ductility.[19] In the steels investigated, cracking was only observed when strain-induced precipitation took place and it may be concluded that loss in ductility during reheating was a necessary condition for cracking to occur.

SUMMARY AND CONCLUSIONS

1. An investigation of susceptibility to heat-affected zone cracking during welding and subsequent heat treatment has been made in three austenitic steels using a welding test.

2. Small hot tears up to 0·03 in. in depth were usually formed during welding at the metal surface in the part of the heat-affected zone immediately adjacent to the weld. The tears were found to be due to liquation in regions of high solute segregation.

3. Extensive intercrystalline cracking occurred during heat treatment after welding. The cracks initiated at hot tears when present and these indeed increased the susceptibility to cracking by providing points of easy initiation. The mechanism of crack formation during heat treatment was quite distinct from that responsible for hot tear formation.

4. Intercrystalline cracking during reheating after welding was always accompanied by the precipitation of a fine intragranular precipitate. In 18Cr–12Ni–1Nb steel the precipitate was NbC, in unstabilized 18Cr–8Ni steel it was $M_{23}C_6$. The steel containing molybdenum which showed little or no intragranular precipitate did not crack.

5. The fine intragranular precipitate was strain induced. Larger amounts of plastic strain gave more dislocations which served as nucleation sites and resulted in finer and more numerous precipitated particles.

6 Increasing the hot strength of the weld metal increased the extent of cracking and produced a finer and more numerous intragranular precipitate.

7. Cracking during heat treatment is due to stiffening of the grains resulting in intercrystalline fracture of a type similar to that observed in creep and hot tensile tests.

ACKNOWLEDGMENTS

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We are very grateful to Mr H. F. Tremlett, Dr J. Nutting, and Dr W. Biggs for helpful discussions, and to Mr C. Humphries who carried out much of the experimental work.

Finally, we would like to thank Dr R. Weck, Director of Research at BWRA for his constant interest and encouragement.

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Iron and Steelworks Engineering
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Investigation of failures in a blooming mill

F. E. Stacey, A.M.I. Mech.E.

SYNOPSIS

During the period 1955–8 a number of serious mechanical failures occurred in a blooming mill, resulting in serious losses of production facilities.

The paper describes the breakages of spindles, drive couplings, and main drive pinion, with the respective metallurgical findings. The final breakage of the mill pinion led to the introduction of tests into the operational mill-loading conditions. These tests were carried out by the field research section of the mill manufacturers, which proved the mill was working well within the conditions indicated under the design specification.

The description of the tests, details of equipment used, and results obtained from the tests are given, with recommendations for standardization of equipment within the industry and recommendations for periodic inspection of such equipment. 1882

THE PRIME PURPOSE of the paper is to describe the reasons for carrying out investigations into the serious mechanical failures that occurred in a rolling mill and to assess mill loading under operational conditions, with the object of minimizing future breakdowns, leading to greater availability for production output.

For this reason detailed calculations have been omitted and the paper has been confined to a description of the principal events.

DESCRIPTION OF PLANT

Figures 1 and 2 show a general view of the mill and the area in which the breakages of plant occurred.

In the case of the top drive spindle, the coupling end at the drive side of the mill housing is forged integral with the shaft; this was the initial design. The bottom spindle, however, had been repaired at some previous time with the aid of a cast-steel coupling.

The main details are as follows:

Type of mill	42 in. blooming mill.
Manufacturer	Davy and United Engineering Co. Ltd.
Commissioned	1950
Details of electrical drive	English Electric Company mill motor, dc, forced ventilated reversing machine having an rms output of 6 650 hp at 62 rev/min. The cut-out torque was 230 metre tons at 62 rev/min. Speed control is by variabl armature voltage from 0 to 62 rev/min and by field weakening from 62 to 150 rev/min. Minimum reversal time not exceeding 1·5 s. The machine was supplied by a Ward–Leonard-controlled Ilgner set.

The motor drives through a coupling, positioned as shown on Fig.1, to a gear unit comprising a pair of double-helical pinions of equal number of teeth. These pinions drive the two spindles, which in turn rotate the operational work rolls.

The mill was installed to cater for an envisaged output of 10500 ingot tons per year. This has greatly increased in the intervening period.

Sizes of material rolled

Blooms 5 in. square up to 9 in. square
Slabs $10\frac{3}{4}$ in. wide plus $4\frac{3}{4}$ in. thick.

Further rolling into either billet or slab sections is carried out on a two-stand Loewy horizontal and vertical roughing mill and a six-stand Morgan finishing mill to form:

Billets	2 in. square to 4 in. square
Slabs	5–16 in. wide by 2–$2\frac{1}{2}$ in. thick
Sheet bar	12 in. wide by 0·25–0·9 in. thick.

Qualities of steel rolled

Carbons up to 1%
Deep stamping and rimming
Silicon and silicon–manganese
Free-cutting
Lead bearing.

Modification to coupling drive

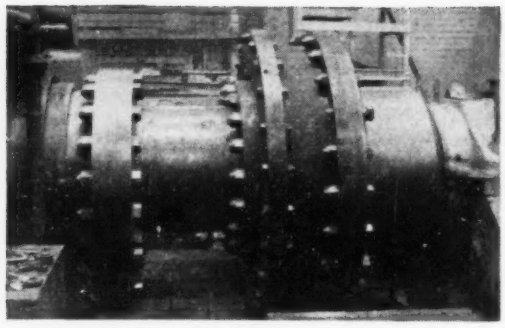
During a period in 1958 a heavy knock was heard at different times in the vicinity of the slab coupling drive. This did not appear to be at regular intervals or at identical times of the rolling cycle.

Upon inspection, no apparent signs of damage could be located, but it was decided that the noise was caused by the slab within the coupling. As a result of this and the occurrence of breakages to coupling slabs, it was decided to investigate the possibility of fitting a flexible coupling to the main drive, replacing the existing slab type.

Good reports had been heard of the Holset coupling when used in similar drive conditions within the industry. This was investigated and a decision made

Manuscript received 5 August 1960.
The author is with John Lysaght's Scunthorpe Works Ltd.

1 *View of section of mill from exit end*



3 *Size 300 Holset flexible coupling fitted August 1959*

to install a size 300 unit with a suitable spacer to meet site requirements, as shown in Fig.3.

This unit was fitted during the stop period of August 1959, an alignment check being carried out at the same time between the motor and gear box units with satisfactory results.

REASONS FOR CARRYING OUT INVESTIGATIONS

During the period March 1955 to September 1958 a number of breakages of a serious nature occurred on the rolling mill, namely (1) the breakage of the knuckle end of a forged-steel top drive spindle, (2) severe fractures to two slab couplings, with visual signs of cracking in a third, and (3) the bursting of the rim sleeve of the top pinion shaft which formed part of the main mill drive. It will be noticed that all the breakages occurred in an area located between the armature drive shaft and the drive side of the mill housings.

It was felt by the management that confirmation should be obtained that the heavy impact loading and reversal conditions taking place on the mill were not exceeding the design conditions, and also that an alignment check should be made between the armature shaft, gearbox, etc.

Metallurgical reports had been received in respect of the broken spindle and slab couplings, which will be mentioned later, but the loss of such plant as pinions necessitated urgent attention.

Breakages of a magnitude such as the latter can cause indirect damage to items such as gearbox casing, pinion bearings, etc.; even if this should cause no further damage to the plant it can, owing to extended deliveries in obtaining replacements, cause heavy loss in terms of money and production facilities, possibly necessitating the closing down of later processes and the inevitable disruption of those processes leading up to the rolling mill.

GENERAL DESCRIPTION AND FINDINGS OF MAIN BREAKAGES

Spindle breakages

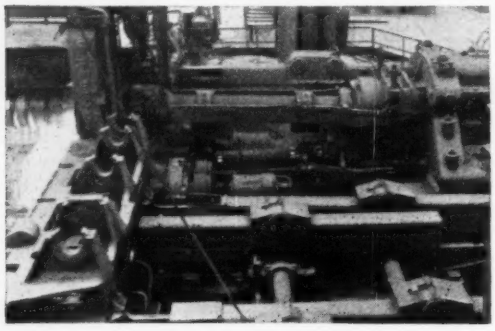
During March 1955, while rolling a silicon quality steel (0·08–0·13%C, 1·5–1·65%Si) of section $12\frac{1}{2}$ in. × 2 in., the top driving spindle suffered irreparable damage at the forged knuckle end which drives the work roll.

A similar spindle, in this case having a cast-steel knuckle end shrunk in position, is shown in Fig.4. The spindle had initially been designed as an integral forging with the driving knuckle end solid to the main shaft, the end driven by the pinions being a cast-steel coupling shrunk into position. The diameter of the replacement cast-steel coupling local to the shaft had been increased to suit the new condition of shrinking in position to the 18-in. dia. spindle.

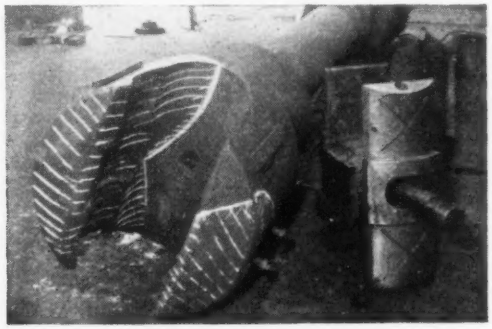
Porosity can be seen on the inner section of the coupling end. This is not an unusual occurrence in castings of this type, where machining allowance may have been minimized or the casting has been thrown to get the best condition from it.

The marking as shown on Fig.4 indicates the extent of the breakage to the spindle, the shaded portions having broken completely away from the main forging.

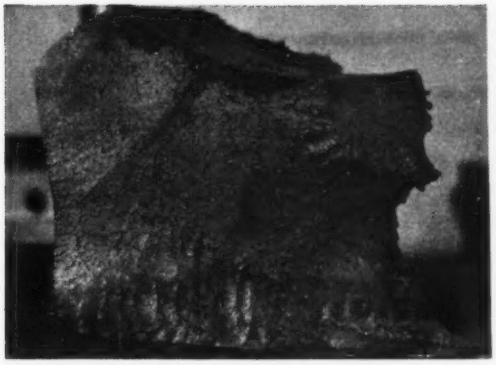
The assembly shown to the right of the spindle is a



2 *View of exit side of mill from an adjacent crane track*



4 *Typical drive spindle and assembly, indicating the extent of breakage*

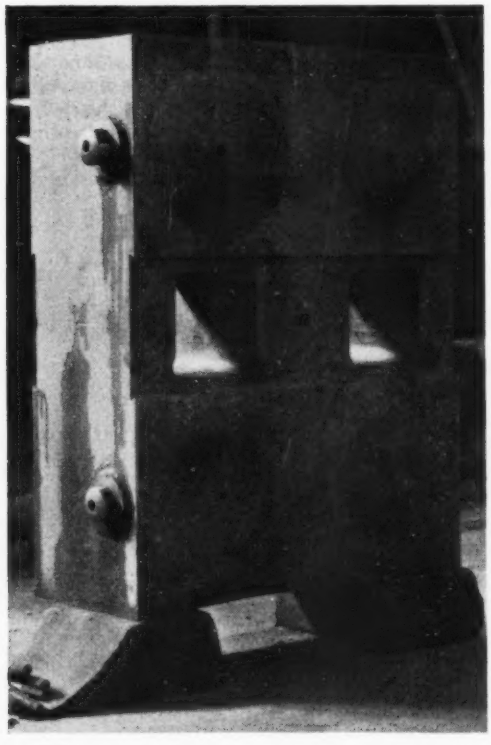


5 *Section of broken drive spindle*

typical unit which fits into the coupling end of the spindle, comprising two phosphor-bronze slipper brasses and a phosphor-bronze split separator block, which in turn holds a central steel block. The bolt as shown is 4 in. dia. and holds the assembly in position, being held through the spindle coupling end.

Visual examination of the fracture (Fig.5) showed the break to be crystalline, with an absence of rubbing, indicating a failure under heavy impact loading. A sulphur print was taken by the metallurgists, who reported a uniform distribution of sulphur with a slight segregation at the spindle slot.

A dark area can be seen at the top, where the machined bore existed. It was found that welding had at some previous time been carried out at this point,



6 *FS drive slab coupling. Dimensions 3 ft 7 in. × 2 ft 10½ in. × 1 ft 2 in. Weight 3 tons*

7 *View showing severity of fracture to slab coupling*

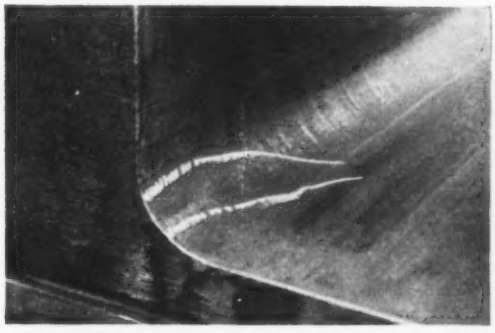
visual signs being apparent under the machining marks.

It was the metallurgists' opinion that slight cracking had occurred at this point, but that despite the presence of this, which could under normal circumstances initiate fatigue failure under the repeated stresses prevailing, this had not occurred.

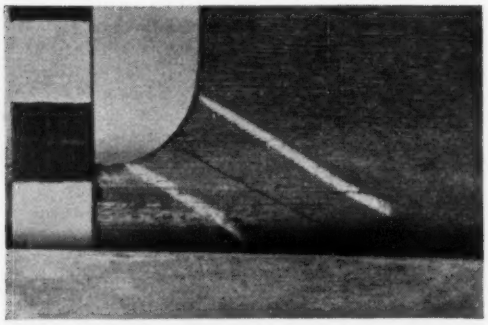
The report also stated that the mechanical properties and structure of the material appeared suitable for the component and that failure was due to heavy impact loading.

Slab coupling failure

Figure 6 is a view of a slab coupling similar to that



8 *View indicating crack in radius and quality of machining to slab coupling*



9 *Slab coupling, showing fracture occurring in radius and on flat face*

which fractured during March 1957, after being used in the rolling of over 3000000 ingot tons.

The fracture occurred diagonally across one of the square holes, through which the coupling bolts fitted into phosphor-bronze blocks, which were located in the coupling.

It will be seen from Fig.7 that failure had taken place through the radii at the corners of the square hole. The structure of the material was fine-grained, with signs of fatigue failure apparent.

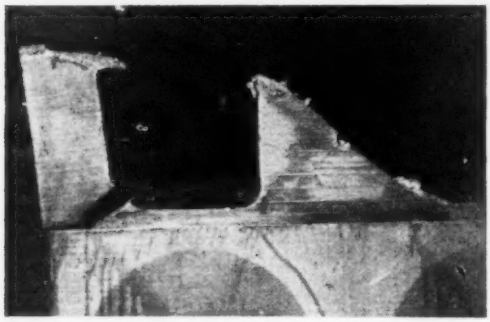
Figures 8 and 9 show the cracking that occurred at the four corners of the hole. Similar signs of cracking were also apparent on the other square hole through which breakage did not occur.

Notice should be taken on Fig.8 of the coarse machining finish with relatively sharp corners caused by blending the square faces to the radius. Two-directional machining must at this point, especially with the quality of the machining, set up a highly stressed component.

The metallurgical report stated that fatigue had occurred symmetrically from the four edges about the centre of the corner radii, the initial cracks being very similar in appearance.

A blue discoloration equivalent to a tempering of bright steel at about 300°C was present at the crack points. Since the coupling was in an area where no heat by radiation from ingots was present, the discoloration was presumably due to internal friction as the crack propagated.

Microexamination revealed a fairly coarse structure, which is usual in a forging of this size. Sulphur printing and etching produced no abnormalities.



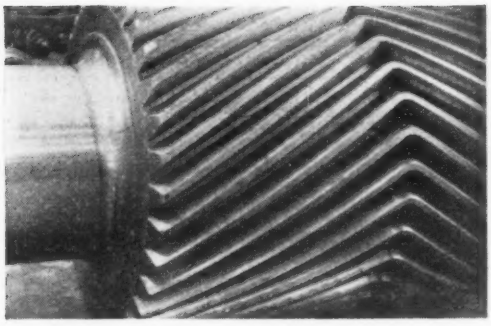
10 *View of fracture in replacement slab coupling*

11 *View of fracture in replacement slab coupling*

It was suggested that benefit would be derived from future slabs if the quality of the finish at the radii was improved. This is considered to be a matter of great importance.

The appearance of the cracks indicated, from previous experience, that they may have been in existence for some time, and the opinion was that the

12 *View of fracture in replacement slab coupling*



13 *Breakage to top mill pinion sleeve*

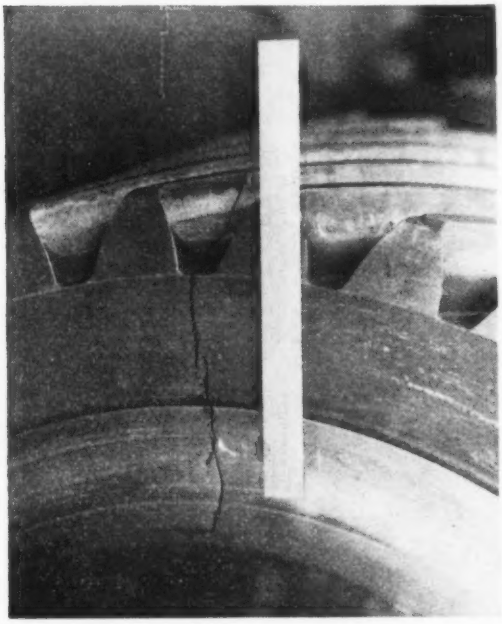
fracture was of a fatigue nature, as seen from Fig.7, the physical and chemical composition being in line with that specified by the manufacturers and designers of the plant.

As might be expected, forgings of this size are not obtainable quickly owing to the manufacturing and heat-treatment requirements. In this instance a steel slab was obtained from a local steelworks, rolled from the largest available ingot; it was machined at Lysaght's works and installed to cover the period required in obtaining a forging of the correct steel analysis.

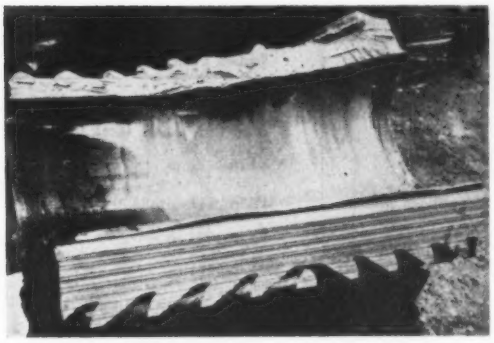
During August 1957, fractures as shown in Figs.10, 11, and 12 occurred in the replacement slab coupling, after being used for only five months. Failure in such a short period was not altogether unexpected in view of the nature of the replacement unit.

The extent of breakage to the slab and the type of fracture are shown in Fig.10. The breakage again originated through the radii at the corners, while other cracks radiate from a central screwhole.

The metallurgical findings were that the chemical



14 *Radial crack through thickness of sleeve in top mill pinion*



15 *View showing method of gear sleeve removal and indicating cracking in tooth root*

analysis varied at two points of the material, which was below the specification of the designers of the mill. This was generally known and accepted for the reasons previously stated, i.e. 0·2–0·24%C as compared with 0·3–0·35%C as recommended.

The coarse end of the material corresponded to the top portion of the ingot as judged by the segregation of the carbon (*see* Fig.11). This coarse structure was lacking in ductility and tensile strength, but the fine structure showed an improvement. A comparison of the structure can be seen on Fig.12.

Samples of the slab were forged from 2 in. to 1 in. square; these showed a great improvement in ductility and Izod readings.

The final comment of the metallurgist was that the slab had been subjected to insufficient hot working to produce suitable grain refinement. It was suggested that future replacements should be produced from larger blooms. Again stress was laid on the need for good-quality machining, though in this instance this was secondary in importance to the unsuitable coarse structure of the material.

The inclusion of this failure is not intended in any sense as criticism of the quality of the material received from the makers, who supplied very quickly indeed, from the largest sized ingot available, thus enabling the production programme to be maintained. It has been included to emphasize the effect of material which has, owing to its initial dimensions, not been subjected to grain refinement by forging and normalizing, which is usual practice in components of this size and application.

Pinion failure

The failure of the top mill pinion in September 1958 was extremely serious, as replacements could, even with extreme pressure being applied, take anything up to 12 months to produce. It was only the generosity of Steel, Peech and Tozer of Sheffield, in permitting the use of an identical pinion that had been replaced and held as an emergency spare, that enabled production output from the rolling mill to be maintained during the period September 1958 to August 1959, when a new pair of pinions became available.

This failure necessitated a great deal of investigation, much more than that required for either the spindle or the slab couplings. This was carried out by representatives of the mill manufacturers. Figures 13 and

TABLE I Typical monthly rolling cycle

Week	Description	Finish dimensions	Tonnage	Cogging dimensions
1	Sheet bar	12 in. wide × 0·25–0·9 in. thick	2500	12½ in. wide × 2½ in. thick
	Blooms or slabs	...	500	5 in. × 5 in. or 11+ in. wide.
	Slabs	5 in. wide × 2 in. thick up to 9 in. × 2 in.	5000	5½+ in. wide × 5 in.
	Billets	4 in. × 4 in.	1000	7 in. × 7 in.
	Billets	2¾ in. × 2¼ in.	2000	5½ in. × 5½ in.
			11000	
2	Sheet bar	12 in. wide × 0·25–0·9 in. thick	2500	12½ in. wide × 2½ in. thick
	Blooms or slabs	...	250	5 in. × 5 in. or slabs 11+ in. wide
	Billets	3¼ in. × 3¼ in.	1050	6 in. × 6 in.
	Billets	2½ in. × 2½ in.	3000	6 in. × 6 in.
	Billets	2¼ in. × 2¼ in.	4200	5½ in. × 5½ in.
			11000	
3	Sheet bar	12 in. wide × 0·25–0·9 in. thick	2500	12½ in. wide × 2½ in. thick
	Blooms or slabs	...	300	5 in. × 5 in. or slabs 11+ in. wide
	Billets	3 in. × 3 in.	3000	5 in. × 5 in.
	Billets	2 in. × 2 in.	600	5 in. × 5 in.
	Billets	2¼ in. × 2¼ in.	4600	5½ in. × 5½ in.
			11000	

14 indicate the horizontal and radial cracking that took place in the sleeve.

As can be seen from Fig.15, the sleeve was removed from the mandrel by planing along the length of the barrel diametrically opposite to the crack. The initial crack in the tooth root, with bedding conditions of the bore to the mandrel, in addition to the sharp corner at each end of the sleeve, can be seen, the latter being a blending radius of the sleeve to the mandrel.

The fact that replacements had to be put to work immediately required a speedy decision to be made as to whether these replacements should be the same as the original gears, (0·5%C steel sleeve shrunk in position on a 0·35%C steel mandrel) or supplied as a solid integral pinion on shaft.

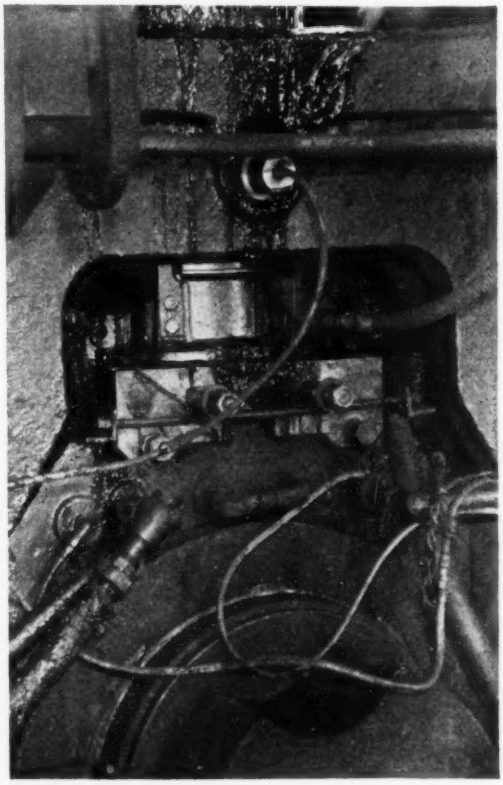
Visual examination indicated that certain teeth on the broken sleeve bore signs of heavy rubbing, accompanied by severe pitting, whereas other teeth showed no signs of wear. It was later found that there had been a slight discrepancy in the gear cutting of the pinions, resulting in a tooth being slightly thinner, followed by insufficient bedding of many other teeth.

It was obvious that the error of gear-cutting would lead to serious overloads on the adjacent teeth as they passed in and out of the mesh, but no definite connexion could be made between the lightly loaded teeth and either the crack which initiated failure or the more serious cracks around the rim.

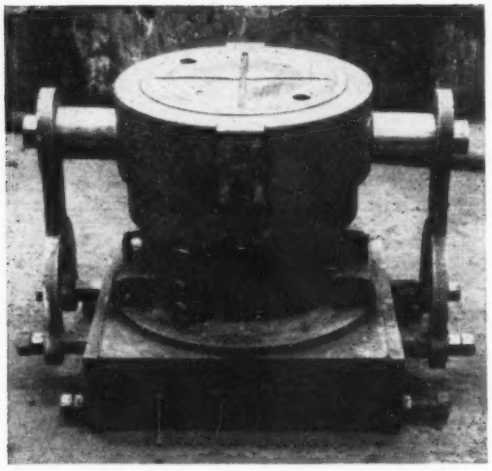
Calculations with a shrink interference, as used with the pinion shaft assembly, determined a hoop stress of 4·96 ton/in² at the root of the tooth. Assuming a full cut-out torque of 7750 ton-in., obtained from two-and-a-half times the mill horsepower of 6650 at 62 rev/min, and also that half this value passed through the teeth mesh, the calculated resultant stress at the root of the tooth due to the bending action of the tooth was 4·06 tons. By the use of a stress concentration and a maldistribution factor to cater for the bad meshing conditions, it was calculated that a total

TABLE II Typical rolling mill schedules

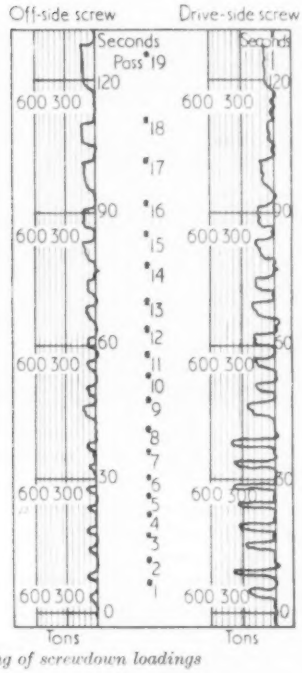
Time	Duration min	Section	Cast no.	Type of steel
9.50	70	5½ in. square	X.15	Si–Mn
11.00	30	5½ in. square	Y.36	Rimming
11.30	60	5½ in. square	X.23	Rimming
12.30	60	5½ in. square	X.22	0·35–0·40%C
1.30	60	5½ in. square	X.21	0·60–0·65%C
2.30	60	5½ in. square	X.24	Rimming
3.30	60	5½ in. square	X.25	0·08–0·13%C killed
4.30	40	5½ in. square	X.28	Rimming
5.10	30	5½ in. square	X.26	SFC lead
5.40	30	5½ in. square	X.28	Rimming
6.10	30	5½ in. square	X.26	SFC lead
6.40	60	5½ in. square	X.29	Rimming
7.40	60	5½ in. square	U.612	0·40–0·45%C
10.15	30	5 in. × 5 in.	W.993	0·20–0·25%C
10.45	60	12½ in. × 4⅜ in.	W.994	Rimming
11.45	30	5 in. × 5 in.	W.993	0·20–0·25%C
12.15	60	12½ in. × 4⅜ in.	W.995	Rimming
1.15	30	9½ in. × 4⅜ in.	Y.33	Rimming
1.45	30	10½ in. × 4⅜ in.	Y.33	Rimming
2.15	60	14½ in. × 4⅜ in.	W.996	Rimming
3.15	60	14½ in. × 4⅜ in.	W.997	Rimming
4.15	30	11⅜ in. × 4⅜ in.	W.998	0·15–0·20%C
4.45	30	11½ in. × 4⅜ in.	W.998	0·15–0·20%C
5.15	60	12½ in. × 4⅜ in.	W.999	Rimming
6.15	25	10¾ in. × 4⅜ in.	W.945	Rimming



16 *Loadmeter* in situ



17 *Loadmeter assembly*



18 *Recording of screwdown loadings*

stress on the tension side of the tooth was equal to 20·5 tons/in² fillet stress.

It was then determined that a fatigue strength of 0·4–0·55 times the UTS would give 18½–25 ton/in². By comparison with the former factored strength, it will be seen that the pinions had been working within this region, particularly if unrecorded shock loads were taken into account.

The final findings indicated that the pinion had failed in service owing to fatigue cracks at the root of the teeth. The analysis showed that the combined stress from bending and shrinking had stressed the material into the fatigue range.

Metallurgical examination showed normal sulphur segregation and distribution, with segregates well clear of the root of the teeth. A micro-section taken from a tensile test-piece showed a fully refined structure of pearlite and ferrite. The pearlite was fairly coarse, but this was not unexpected owing to the retarded cooling at the mid-wall position of the rim.

Guided by the results of the investigation and the time factor in obtaining hollow forgings, it was decided to manufacture the replacements as integral pinions with shaft, in material of En.9 specification. The original pinions were to be resleeved later with material also to En.9 specification, the interference between shaft and sleeve being reduced to 0·020 in. from 0·026 in. (i.e. 0·001 in. per in. dia.).

PRELIMINARY ARRANGEMENTS FOR TESTS

During October 1958 a meeting was arranged with the manufacturers of the mill, who agreed that tests should be carried out by their personnel upon the plant concerned to determine mill loading under operational conditions.

At this time their firm was developing a new type of loadmeter suitable for installing on rolling mills to determine operational screwdown loading. It was anticipated that a pair of units suitable for loads of up to 1500 tons each would soon be available, but as a great deal of preparation was necessary to work out the anticipated programme and to manufacture extra fittings to suit the conditions existing on this particular plant, theoretical calculations would be made to determine mill loading from operational conditions. This is mentioned to clarify the reason for the delay in carrying out the tests.

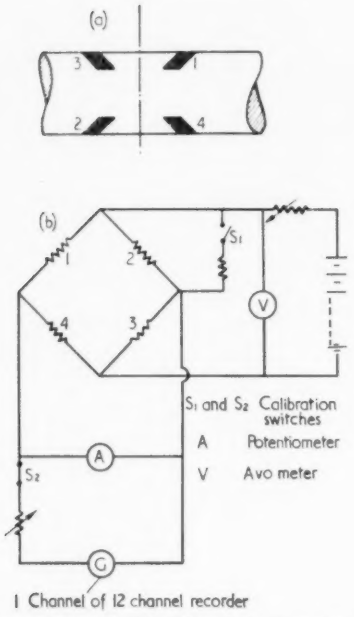
It was suggested that the object of the tests should be to determine the mill loading conditions, the torque applied to the driving spindles, operating temperature, mill power consumption, and bearing pressure conditions of the motor shaft, and generally to determine if the operating conditions were covered by the design of the mill.

When carrying out the proposed tests, which would of necessity be at reduced drafting with additional passes, owing to the use of the second-hand pinion (a practice which is, under normal circumstances, to be deplored) it was decided that, after the flexible coupling had been installed, identical tests should be carried out with the same conditions as those existing for the original slab type of coupling drive, thus obtaining information about the relative performance of the mill under the two types of drive.

It was further suggested that an additional test should be carried out with drafting and number of passes simulating the conditions existing at the time of the pinion breakage.

With these points in mind, it was felt that even if the proposed tests proved negative, they would not have been wasted. From Lysaght's point of view, a great deal of satisfaction could be obtained from the knowledge that the mill was not being overloaded; it might even give information that would be beneficial in modifying their rolling sequence, and that from the research angle facilities would be made available upon an actual operational basis for determining conditions existing upon a specific mill, thus providing data for use in future mill design.

It was fully appreciated that, owing to the varying sections and qualities of steel being rolled within any shift, it would not be necessary for the research team



a Arrangement of Saunders-Roe foil torque gauges for each spindle
b circuit diagram from strain gauges to galvanometer recorder

19 *Torque loadings*

to be in attendance full time, but rather that tests should be selected over the actual rolling cycle, which covered a four weekly period.

CONDUCT OF TESTS

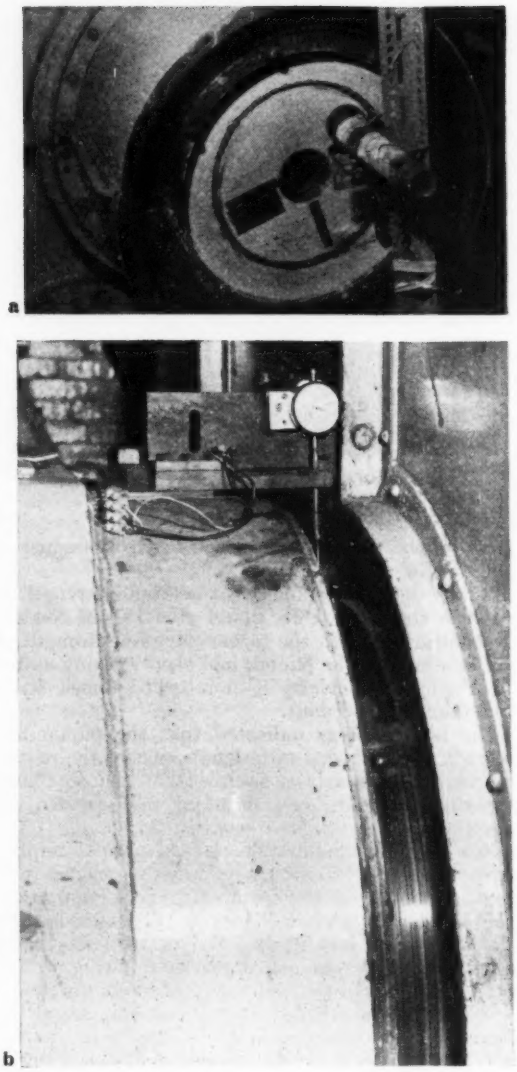
To enable the research team to select the periods during which greatest benefit could be gained without necessitating full-time attendance, typical monthly rolling cycles and daily production programmes were made available (Tables I and II). This enabled them to select the period of the week in which to be in attendance over the cycle, and further enabled the team to carry out selective tests during the time they were in attendance.

Tests began on 1 June 1959 and were carried out at selected times until 12 September. During this time the mill had been closed down for two weeks to facilitate modification of the existing motor shaft necessary for fitting of the new coupling and installation of new pinions and bearings, in addition to the usual routine and annual maintenance programme.

INSTRUMENTATION FOR TESTS

Two loadmeters, each of 1500-tons capacity, were fitted between the mill screws and the top-roll bearing chock to determine the operational load on each screw (Fig.16). Each loadmeter consisted of a steel column, on to which were bonded eight electrical resistance strain gauges, four mounted vertically and four horizontally. The gauges were connected in a Wheatstone bridge circuit, each arm consisting of two gauges. The whole assembly was installed within a casing, as shown in Fig.17.

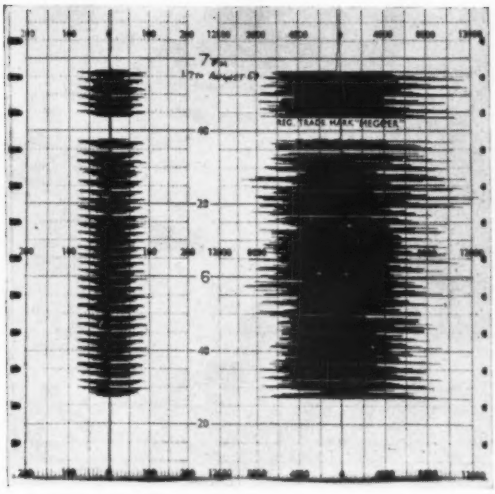
When a load was applied, the vertical strain gauges were in compression and the horizontal gauges in tension. The output from the bridge, which was proportional to the load being applied, was recorded separately on a two-channel recorder; a record of an actual condition is shown in Fig.18.



20 (a) *Mill speed and motor shaft end float instrument assembly*, (b) *strain gauge and visual clock gauge to determine lift of shaft at inboard bearing*

Electrical resistance strain gauges were fitted midway along the length of the top and bottom spindles in the plane of principal stress to determine spindle torque. These were arranged in the form of a temperature-compensated Wheatstone network sensitive only to torsional strain, the output being fed to the 12-channel recorder as two traces. Figure 19 indicates the arrangement of the torque gauges and the circuit used.

In addition to this, the mill speed and ampere readings were recorded upon a two-pen recorder unit situated in the mill motor room; this is a continuous process when the plant is operating. Figure 20 shows a recording of the mill speed in both directions with ampere readings for a corresponding period.

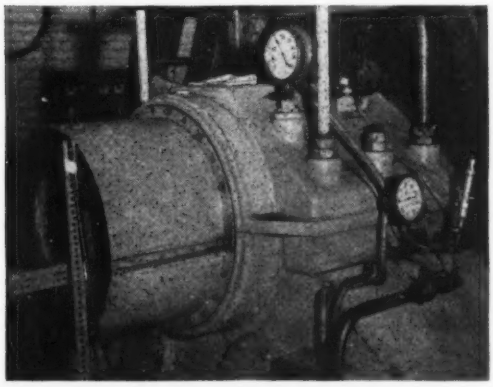


21 *Ampere and tachogenerator recordings*

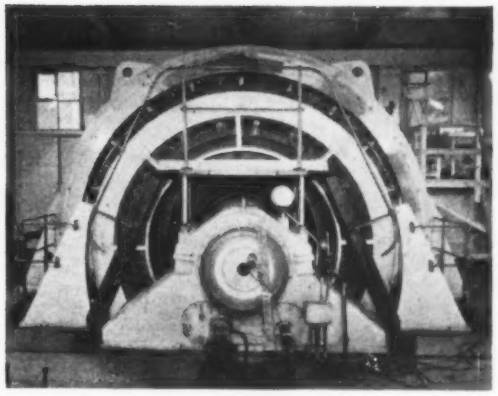
Pressure transducers employing electrical resistance strain gauges were screwed in existing tapped holes on both the motor bearing housings. The operational oil pressures of the bearings were recorded as two traces on a 12-channel recorder.

To determine the horizontal and vertical movement of the armature shaft, displacement transducers were fitted. These were spring steel strips fixed at one end, with the free end in contact with the end of the shaft or its diameter for horizontal or vertical displacement respectively, strain gauges being fitted at the end of the strip. The strain gauge for horizontal displacement can be seen on Fig.20*a* to the right of the photocell arrangement. The circular marking on the shaft end was caused by the rotation of the shaft against the ball end of the displacement transducer. A visual aid in the form of a dial clock gauge was fitted to indicate if any radial movement of the shaft was present, for reference purposes only (Fig.20*b*). Impulses from the transducers were recorded as two separate traces on the 12-channel recorder.

To determine the temperature of the ingots during the complete rolling sequence, recordings were made with the aid of hand surface pyrometers, when the ingot was placed on the entry roller table and after leaving the final pass. The temperature before each pass through the mill in a forward direction was recorded by a photocell pyrometer positioned on the offside entry manipulator. This temperature was also recorded upon the 12-channel recorder.



22 *View showing pressure transducers* in situ



23 *General view of motor room layout*

Figure 23 shows the various items of equipment used, with the operations deck at the rear right-hand side overlooking the mill pulpit; Fig.24 shows the 12-channel recorder unit and electrical units.

During the period of each test, information was recorded, with the aid of a portable tape recorder, of operational mill draft and clock readings, the time during which the material was in each pass, the time taken in transferring material from one pass to the next, and the total period taken in rolling the complete ingot.

After completion of rolling and before the material left the run-out table, the thickness was measured with the aid of calipers to enable a mill spring curve to be determined (Fig.26). This was later to be used in conjunction with an estimated spread in determining the entry size of the ingots into the respective pass.

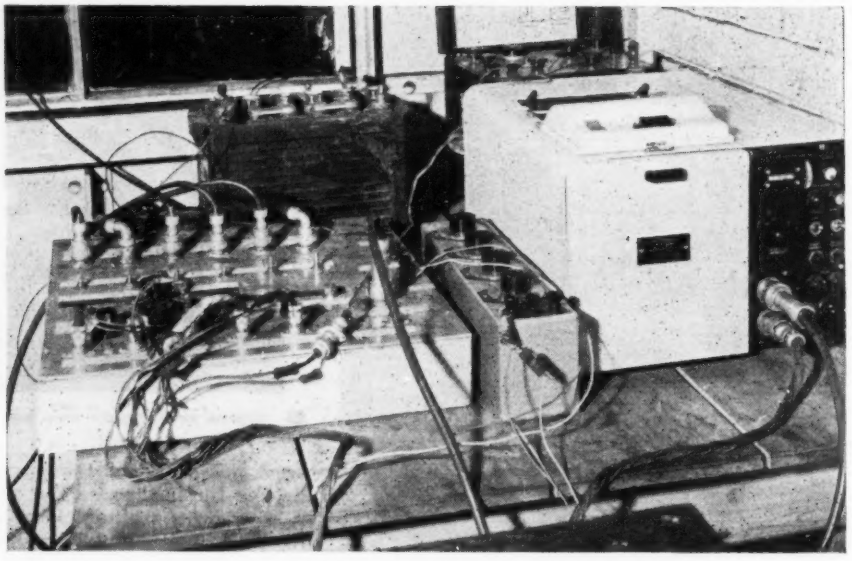
Comments on the filmed results

Figure 25 is a small portion of the full recording, and possesses little value until the calibration data are determined for each and every trace. The designation to the left of the chart is self-explanatory, the numbers 1 to 5 indicating the conditions at the time when the ingot material was in transit through the work rollers in either direction, i.e. odd numbers in the forward and even numbers in the reverse directions.

Calculated and tabulated results

Table III gives the results of the conditions prevailing on the mill at the time of rolling a specific billet. The fact that each test required a similar table of results and recordings indicates the complexity and quantity of the work carried out by the research team.

The headings in general are self-explanatory, but a few comments may be helpful. Column 1 indicates the number of the pass relative to the tabulated results. The details given under column 2 represent the calculated thickness of the material on entering the rolls, with an allowance included to accommodate the spread of material that had occurred during the



24 *Layout of recording instruments*

previous pass plus an amount taken from the mill spring curve (Fig.26), which caters for the experimentally determined spring of the mill, e.g. elongation that may take place at the housings due to screwdown loading, clearances in bearings, screwdown nuts, etc.

The following column (3) indicates exit dimensions of the ingot after a pass and is equivalent to the next entry condition, the exception being when the ingot is turned. Column 4 (H–h) is therefore equivalent to the reduction of the material within any specific pass.

Information given under column 5 is as taken from the screwdown load chart (*see* Fig.18 for typical recording).

Column 6 is shown under four sections, denoting the torque applied to the top and bottom spindles, the total torque applied to the mill, and the ratio of applied torque to the spindles. During the early passes the bottom spindle carries the greater torque, this being transferred to the top unit in the later stage.

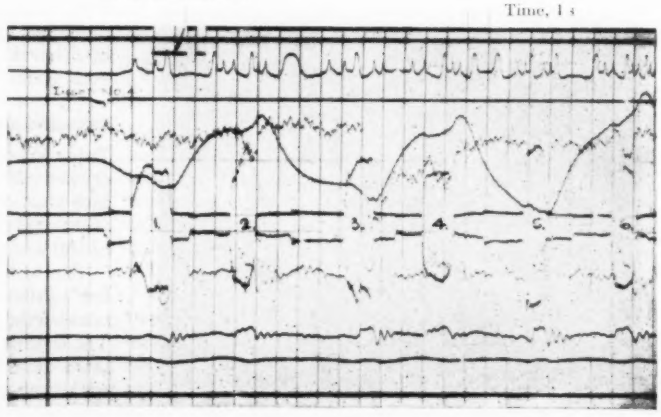
The motor speed in revolutions per minute, as taken from the tachogenerator and calibrated with the photocell recorder, is shown under column 7. These speeds were confirmed by the relative motor room chart, a typical recording being shown in Fig.21.

The time during which the material was in the pass was important for calculation purposes; typical readings are shown under column 8.

The mill motor loading amperes and voltage are recorded under columns 9 and 10 respectively. Details of the horsepower of the mill at the various pass conditions, as calculated from information gained from the tests, are shown under column 11, with a corrected value under column 12 where a mill efficiency factor has been taken into account. It had been assumed that the mill efficiency for these conditions should be 90%, but this is rather conservative, as it may at times be up to 95% efficient.

At times the evaluated horsepowers are in excess of those stated in the mill specification. All the values as given under column 12 require conversion into terms of r.m.s. horsepower, the average value of the specific test being shown at the lower portion of Table III.

Details of recorded temperatures before and after completion of rolling are included upon the chart. A


Ingot identification
Time, 1 s
Motor speed, from photo electric cell
Ingot temperature, from pyrometer on manipulator
End float of motor shaft
Motor speed, from mill tachometer
Bottom spindle torque
Top spindle torque
Vertical displacement of shaft
Motor bearing pressure, inboard
Motor bearing pressure, outboard

25 *Section of filmed recordings*

TABLE III Conditions at mill

Temperature of ingot before rolling = 1190°C
Temperature after rolling = 1140°C

Total in roll time = 52·9 s
Time to roll ingot = 141 s

1	2	3	4	5			6				7	8	9	10	11	12
				Measured load, tons			Measured torque, ton-in.									
Pass no.	*H*	*h*	δ	Drive side	Off side	Total	Top spindle	Bottom spindle	Ratio top/bottom	Total	Mill speed, rev/min	In-roll time, s	Current, A	Voltage	Horsepower (from torque and rev/min)	Horsepower (assuming 90% mill efficiency)
1	24	22·54	1·46	352	90	442	1210	1430	0·85	2640	23·5	2	3600	660	2210	2460
2	22·54	20·71	1·83	457	105	562	1690	2000	0·88	3690	38·2	1·2	6200	870	4750	5300
3	24·80	22·37	1·43	405	97	502	1630	1990	0·86	3530	44·1	1·2	4700	795	5590	6110
4	22·37	20·56	1·81	375	82	457	1160	1480	0·78	2640	44·1	1	4200	795	4150	4610
5	20·56	18·69	1·87	420	120	540	1630	1660	0·98	3290	44·1	1·3	4800	810	5150	5730
6	18·69	16·75	1·94	420	97	517	1590	1720	0·93	3310	41·2	1·4	4600	810	4850	5400
7	16·75	14·6	2·15	442	112	554	1790	1830	0·97	3620	47·0	1·5	5600	840	6020	6700
8	14·6	12·65	1·95	420	105	525	1680	1660	1·01	3340	41·2	2	5000	870	4900	5430
9	23·13	18·58	4·55	300	172	472	2380	2120	1·12	4500	35·3	2·5	6000	855	5650	6290
10	18·58	16·35	2·23	240	135	375	1260	1140	1·10	2400	60·0	2	4000	960	4250	4720
11	16·35	14·49	1·86	225	120	345	1170	1140	0·94	2310	44·1	2·8	3400	840	3460	3850
12	14·49	12·4	2·09	232	120	352	1310	1140	1·15	2450	52·9	2·2	4800	1050	4600	5110
13	14·02	11·31	2·71	240	120	360	1310	1140	1·15	2450	44·1	3·4	4300	990	3840	4270
14	11·31	9·35	1·96	210	105	315	940	1030	0·91	1970	58·8	2·8	4000	1050	4180	4650
15	12·97	9·97	3·00	180	150	330	1400	1140	1·22	2540	64·7	3·4	5100	1020	5840	6500
16	9·97	7·58	2·39	165	135	300	1170	920	1·27	2070	64·7	4·2	4600	1080	3960	4400
17	9·86	7·26	2·60	120	135	255	1070	800	1·34	1870	61·7	5·4	4200	1110	4100	4550
18	7·26	5·17	2·09	52	142	194	560	570	0·98	1130	79·4	4·4	3900	1140	3180	3530
19	7·94	5·55	2·39	97	150	247	890	760	1·19	1650	73·5	8·2	3700	1200	4260	4740

r.m.s. value of horsepower during this schedule = 4370

26 *Graph showing mill spring: loading at screwdown. Typical example: mill spring at 350 tons = 0·4 in.*

temperature drop occurred during the complete sequence of 50°C.

Findings from tests

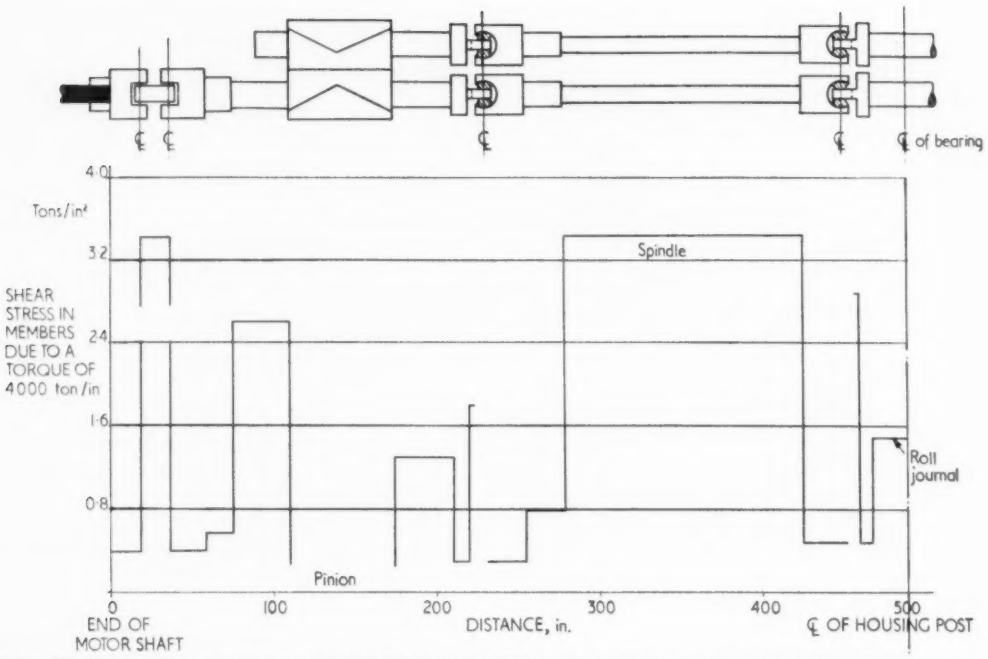
The knock within the coupling area occurred 2 s before the entry of the material into the ninth pass, the slab being at that time in a vertical position. This corresponded to the point of rolling, where a heavy drafting occurred after the material was turned on its edge owing to spreading having taken place. The vibration caused by this knock occurred only in the vicinity of the coupling and gearbox, showing on the torque trace as oscillations.

Disappearance since the fitting of the flexible coupling indicated that the noise emanated from the slab coupling but did not verify that the cause had been removed. The absence of unusual values of loading or spindle torque, however, indicate that the process of rolling did not inflict any high stresses upon the plant.

Inspection carried out during the stop period to the pinions, palm end of spindles, and carrier pivots showed no indication that they had been subjected to heavy impulse loads. The pinions did, however, show signs of excessive pitting and wear, the result of using an unmatched second-hand pinion that was in a relatively poor condition when fitted. The original bottom pinion at the time of the breakage had been in good condition, but had suffered greatly in simulating the mating conditions and irregularities of the temporary replacement.

The recorded trace of the armature shaft movement showed up as a mirror trace of the offside bearing pressure, the latter increasing as the shaft floated outwards. The maximum recorded float was 0·012 in., corresponding to a maximum pressure fluctuation of 33 lb/in². The radial movement or lift on the shaft at the inboard bearing was found to be 0·005 in.

Maximum loading and torque occurred during tests carried out in the rolling of what is considered to be



27 *Stress distribution through mill train for spindle torque of 4000 ton/in. from tests carried out excluding pinions*

the most difficult steel within Lysaght's manufacturing range, silicon–manganese. The maximum load recorded was 700 tons, compared with a permissible load of 1 ton/in² of mill post area, a total of 1800 tons. The greatest torque recorded was 4000 ton-in., calculations showing that at this torque the stresses in the slab coupling, pinions, spindles, and roll necks were well within the design limits of the respective materials.

When simulating the conditions existing at the time of the pinion breakage (15 passes), it was noticed that the torque recorded on the first eight passes exceeded those of the revised drafting conditions (19 passes). This was found to be in the region of +33% when rolling a $5\frac{1}{2}$ in.×$5\frac{1}{2}$ in. billet section. An estimated increase of 32% would be applicable for a $12\frac{1}{2}$ in.× 2 in. section. In no instance, however, did the total torque exceed 4000 ton-in.

Under all conditions of the tests carried out, the motor was found to be working well within its capacity. During the rolling of a $5\frac{1}{2}$ in.×$5\frac{1}{2}$ in. section, the approximate r.m.s. horsepower was 4000 with a 15-pass cycle, whereas with a 19-pass rolling the r.m.s. horsepower was 4500, the time factors being very similar.

GENERAL CONCLUSIONS

From the results of the tests carried out and calculations made, it was proved that the shear stress in the members of the mill train were not exceeding conditions envisaged and covered in the initial mill design (Fig.27 shows the distribution of stress through the above). If the mill is operated under conditions prevailing at the time of the tests, major breakdowns to equipment should not occur, as the plant is working well within its capacity.

The modifications carried out by the fitting of the flexible coupling indicated no appreciable effect upon the performance of the mill, but they dispelled the unexplained noise at the coupling. They should result in the elimination of maintenance troubles in this area, and remove the mechanical shock from the motor, which is absorbed by the rubber elements in the coupling.

The results show that the 15-pass rolling cycle could be adopted for general working during the rolling of $5\frac{1}{2}$ in.×$5\frac{1}{2}$ in., 7 in.×7 in., and $12\frac{1}{2}$ in.×2 in. sections, resulting in a possible increase of tonnage output of up to 10%.

The results and information gained from the tests should prove most useful in future design of rolling mills and follow the pattern of conditions envisaged and allowed for in design specifications. Above all, the tests provided facilities for gaining information which had previously not been available, and which should in the future prove most beneficial.

To decide whether to hold spares of the nature of pinions, etc., to await installing upon a breakage or life basis is not the object of this paper. However, the author believes that, in industry as in life, one must take a certain number of risks. In the instance of the pinions, if it had been decided to take a calculated risk, it was unfortunate that it failed, but by good fortune and good will between competitors, it was possible to borrow a second-hand pinion that had been replaced on a life and wear basis. The benefit to be obtained from standardization of major pieces of plant could not be better illustrated.

It is fully realized that many people will feel that spares of this nature should be available. To this the author would agree, since on the price argument, which it must be agreed can be considerable, no firm should place an essential part of its plant in jeopardy,

when such components may take up to 12 months to obtain as replacements.

Arising from the pinion breakage two factors are worth consideration. Occasionally, and particularly at shop-floor level, past experience and rule-of-thumb methods are used for the determination of shrink and press fit tolerances. These can possibly result, but not necessarily immediately, in breakage by hoop stress, etc. At times insufficient bedding of teeth is carried out, followed by periodic inspection, with if necessary further bedding. This may not always be the fault of the engineering staff, but may be necessitated by production requirements. It is suggested that an extra period of, say, 12 h may result in a later saving of two to three days' downtime and loss of production.

Periodic inspection of gears, etc. should not be ignored just because they are encased within a gearbox. These need the same care and attention as any other part of the plant. As has been Lysaght's practice, an annual inspection and crack detection of teeth and journals, with the aid of magnetic ink or some other method, is considered essential.

ACKNOWLEDGMENTS

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Some aspects of lighting in the iron and steel industry

D. R. M. Nisbet

SYNOPSIS

Because of the variety of functions associated with the iron and steel industry, the lighting requirements vary greatly. Where statutory regulations have to be met, ample guidance is available to establish standards, but in most areas there is a qualified exemption from statutory requirements. In these areas the standards have to be decided by the individual lighting designer, who in establishing his standards should be guided by current practice tempered by past experience and local standards. By careful consideration of the forms of lighting now available, by assessing economics on a full-life basis, and by arranging that maintenance is of a high order, good lighting can be provided at a justifiable cost. 1891

MUCH has been and is still being written on the subject of lighting. This material is produced, in most cases, by those who may not be strictly classed as users. It is hoped in this paper to present a user's view on lighting, or on some of the factors that may influence the user, a large integrated iron and steelworks, when formulating a lighting policy.

It is most important to decide first of all how high a standard of lighting is to be adopted. No one would disagree today that all lighting must be good lighting, good for the conditions for which it is provided. However, good lighting in one area, say a melting shop, might be totally inadequate for the needs of a machine shop, and conversely the intensity required in a machine shop will, most likely, be in excess of what is necessary in a melting shop. It follows, therefore, that each function or working area must be considered separately, and the level of illumination required for each must be decided on its own merits.

It is most important that, having decided on a value of lighting intensity for any particular area, this value should be maintained at all times. It is just as important to maintain an established standard of lighting as it is to maintain, say, a standard voltage or a standard oil pressure, in fact, any standard that has been logically established on the works; to do otherwise is wasteful.

To assist in establishing the correct standard of lighting to be employed under the varying conditions that exist on a large iron- and steelworks, there are several published references that can be consulted. In fact, certain circumstances demand a minimum standard of lighting that has to be provided by statute. Where these minimum standards apply in an iron and steelworks depends to some extent on the interpretation placed on the appropriate regulation, for although there is a qualified exemption given to some parts of iron- and steelworks, the regulation must apply in others. A reminder of these regulations might not be out of place at this point.

Regulation 2

(*a*) The general illumination over those interior parts of the factory where persons are regularly employed shall not be less than 6-ft candles measured in the horizontal plane at a level of 3-ft above the floor. Provided that in any such parts in which the mounting height of the light sources for general illumination necessarily exceeds 25-ft measured from the floor or where the structure of the room or the position or construction of the fixed machinery or plant prevents the uniform attainment of this standard, the general illumination at the said level shall not be less than 2-ft candles, and where work is actually being done the illumination shall not be less than 6-ft candles, or the greatest reasonably practicable illumination below 6-ft candles.

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The author is with the Appleby–Frodingham Steel Company.

(*b*) The illumination over all other interior parts of the factory over which persons employed pass shall when and where a person is passing be not less than 0·5-ft candles measured at floor level.

(*c*) The standard specified in this regulation shall be without prejudice to the provision of any additional illumination required to render the lighting sufficient and suitable for nature of the work.

It is clear from this regulation that although minimum figures of 6 lumens/ft^2 and 2 lumens/ft^2 are quoted, in certain circumstances these would be considered inadequate. In other words, the maximum lighting intensity necessary in each case is left to the discretion of the lighting designer. It would seem, therefore, that in those places in an iron and steelworks where regulation 2(a) applies (e.g. in workshops) a recommended code of practice should be adopted.

As previously stated, certain exemptions are allowed, and these are concerned with the type of work being done on the plant. Details of the exemptions are given in subsection 5(b) of the regulations. Here it is found that regulation 2(a) will not apply to 'parts of factories in which the following are carried out. The conversion of iron into steel, the smelting of iron-ore, iron or steel rolling, hot rolling or forging, tempering or annealing of metals'.

From this it follows that, whereas there are certain working areas where, being bound by statute, guidance is given by authority on the minimum standard of lighting to be provided. There are other working areas where no such guidance is given. It is with these areas that the remainder of the contents of this paper are largely concerned. Examples of these are melting shops, cast houses, rolling mill buildings, etc.

How then is a figure for a satisfactory standard of lighting to be assessed for these areas? The first question that arises is, why are such areas exempt from statutory regulations? Presumably, one of the main reasons is because it is recognized by authority that the actual area where work is being done is only a small percentage of the total area that has to be lit. Another reason may be that the class of work normally done in these areas is such that a high intensity of lighting is not necessary to enable the performance of the work to be carried out satisfactorily. It is, therefore, unnecessary and costly to overlight such large areas.

On deciding a suitable illumination level for these areas, there are several factors which ought to be given due consideration. Some of these factors are obvious, and some are not quite so obvious, but still important. Four such factors are:

1 Environment and safety
2 Experience and practice
3 Cost
4 Maintenance

ENVIRONMENT AND SAFETY

The importance of environment has already been indicated by stating that all lighting should be good lighting. It will not be good lighting if any physiological strain is imposed on personnel due to limitations in the design of the installation. However, it is not necessary to discuss this aspect further here as this discussion would merely be a repetition of what has been said many times before by several authorities, who give ample guidance on the various factors to be taken into consideration.

It goes without saying that for safety's sake, too, all lighting should be good. However, safety ought to be an end rather than a means to good lighting. Any lighting system which fails to satisfy the needs of safety is an indictment and could be construed as a breach of regulations.

EXPERIENCE AND PRACTICE

On any iron- and steelworks there are still many buildings in which the lighting intensity is of the order of 1–2 lumens/ft^2, and in these buildings work has been going on satisfactorily for many years. Production records have been broken with monotonous regularity and it is doubtful if the most plausible advocate of better lighting could prove that an increase in lighting intensity would increase production. Without condoning such arrangements, it is obvious that certain work can carry on under such conditions. This fact is significant as it provides a theoretical bottom limit to the standard of lighting required and it provides some sort of yardstick to use when consideration is being given to the top limit.

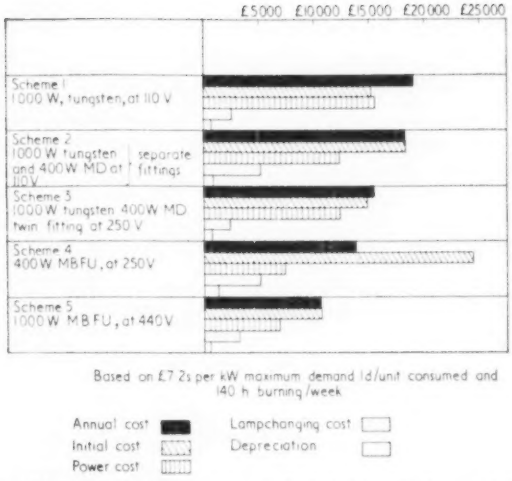
While, as has been stated above, work can be carried on satisfactorily in comparatively low lighting intensities, it is now accepted that lighting standards of yesterday are well below what are considered as normal today. This applies not only throughout industry but even in domestic surroundings. Lighting is an amenity and the demand for better lighting has grown along with the demand for other amenities. It follows then that something more than just adequate lighting has to be provided to keep abreast of the times.

The snowball effect of improved lighting has to be borne in mind. It has already been stated that a good standard of lighting has to be provided by statute in some parts of the works. Immediately this is done, all other installations which are not lit to these standards become badly lit by comparison. However, there is no justification for improving lighting beyond the required intensity in order to preserve uniformity of values. A melting shop does not need to be lit to the same intensity as a machine shop or, to take it to extremes, a drawing office.

COST

Obviously, one of the most important factors to be considered when a lighting scheme is being designed is the cost. But for cost, all areas could be lit to near daylight conditions. This might please the lighting engineer, but certainly not the accountant. It is fair to state, therefore, that the amount of light to be provided in any working area is somewhat of a compromise between the best lighting possible and the most economic. Note, however, that the most economical lighting scheme will seldom be the cheapest installation. Many considerations enter into the economics of lighting and not least of these may be efficiency of production. Claims that good lighting can increase productivity are continually being made. The standard of lighting, therefore, should be the best justifiable.

In order to assess the various factors affecting the cost of lighting, a typical area (an open-hearth melting shop) has been considered. For comparative purposes, the costs for a number of different schemes of lighting


£5000 £10000 £15000 £20000 £25000
Scheme 1
1000 W, tungsten, at 110 V
Scheme 2
1000 W tungsten and 400 W MD at 110 V
separate fittings
Scheme 3
1000 W tungsten 400 W MD twin fitting at 250 V
Scheme 4
400 W MBFU, at 250 V
Scheme 5
1000 W MBFU, at 440 V
Based on £7 2s per kW maximum demand 1d/unit consumed and 140 h burning/week
Annual cost
Initial cost
Power cost
Lampchanging cost
Depreciation

1 *Diagram showing annual cost of selected installations and the elements that make up the annual cost*

have been estimated. For simplicity the comparison has been made on a basis of the same total lumen output which, while not strictly accurate, is near enough for these purposes.

The systems that have been compared are:

Scheme 1 1000-W tungsten-filament lamps only, voltage 110 V

Scheme 2 1000-W tungsten-filament and 400-W mercury discharge lamps, voltage 110 V (separate fittings)

Scheme 3 1000-W tungsten-filament and 400-W mercury discharge lamps, voltage 250 V (twin fittings)

Scheme 4 400-W M.B.F./U., voltage 250 V

Scheme 5 1000-W M.B.F./U., voltage 440 V.

In each case, costs have been estimated for:

(*a*) Initial installation (including the cost of fittings, transformers and wiring, lamps, and labour to install)

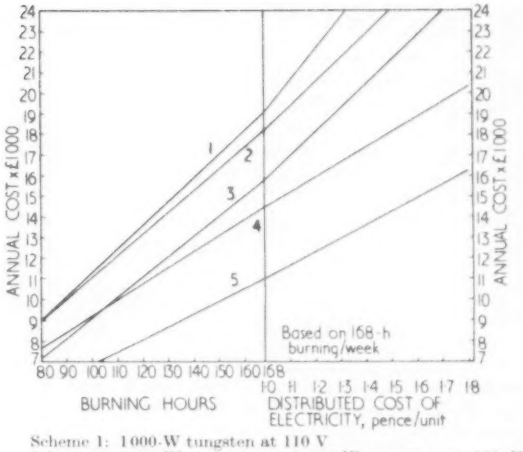
(*b*) power consumption

(*c*) Lamp changing (including the labour costs involved).

Note: In all cases list prices have been used.

The life of the fittings has been given as 15 years and the life of the installation as 30 years. The number of burning hours per week has been taken as 140 and the cost of power as £7 2*s*. per kW maximum demand plus 1*d*. per unit used. From these figures, and taking depreciation on a straight-line basis, annual costs have been calculated and a block diagram prepared (Fig.1). This shows the comparative annual costs of the various forms of lighting and the main elements making up these costs. Curves have also been drawn to show the effect on annual cost of a variation in distributed power cost and also of a variation in burning hours on the several lighting schemes (Fig.2).

The important part of Fig.1 is the annual cost, and it is interesting to note, on examining the breakdown of this cost, that the item which most influences the annual cost is the cost of power. It goes without saying, therefore, that the system which consumes least power has a distinct advantage over all other systems. However, it must not be assumed that this advantage outweighs all others. Before a choice of system is made, there are several aspects to consider. The most important of these are now discussed under their several headings.



Scheme 1: 1000-W tungsten at 110 V
Scheme 2: 1000-W tungsten and 400-W mercury at 110 V (separate fittings)
Scheme 3: 1000-W tungsten and 400-W mercury at 250 V (twin fittings)
Scheme 4: 400-W M.B.F.U. at 250 V
Scheme 5: 1000-W M.B.F.U. at 440 V
Based on distribution cost of 1*d*. per unit and 168 h burning per week

2 *Comparison of annual cost for varying burning hours and power cost*

Application

One of the essential features of any lighting scheme is that there must be an even distribution of the light over the area. In order to achieve an even distribution of light, the ratio of the number of lighting fittings to the area must be the correct one. For example, one high-wattage lamp in the centre of a large room may produce a total amount of light which is sufficient to light the room adequately, but most of the light will be cast under the lamp itself while the extremities of the room will be comparatively poorly lit. This principle applied to a large building would give a 'spotty' distribution which would be most undesirable. To achieve a satisfactory distribution there is a minimum number of fittings which must be used, and this minimum is largely dependent on the design of the fitting and the height at which the fitting is mounted. It follows then that lamps having a very high lumen output can only be used in comparatively high buildings. The lamps that are shown in scheme 5, for example, are very high output lamps, having about three times the lumen output of the same wattage tungsten-filament lamp of scheme 1. This means that it would only be practicable to use this type of lamp in a very high building. The lamps quoted in the other schemes have a much more general application.

Working voltage

There are three standard voltages in general use for lighting: (i) 110 V, (ii) 250 V, and (iii) 440 V.

110-V system

110 V is fairly widely used in heavy industry for lighting. The 110 V can be obtained through double-wound transformers which have the centre-point of the secondary winding connected to earth. This ensures that the voltage, above earth potential is only 55 V, a voltage generally considered to be safe. An added advantage of this system is that a convenient

safe supply is made available for socket outlets for portable tools.

One of the advantages of tungsten-filament lighting is that the installation can be operated satisfactorily at 110 V. In fact, 110-V tungsten-filament lamps are more efficient than their 250-V counterparts because of the higher light output per watt which the lower-voltage lamps give. However, as shown in Fig.1, tungsten lighting is costly in power consumption.

One way in which the high power costs can be minimized is to use mercury-discharge lamps in conjunction with the tungsten lamps. This arrangement has come to be known as a 'blended' system of lighting and the effect of this on cost is shown in Fig.1. There is, however, one disadvantage with this system, namely that the life of a 110-V mercury-discharge lamp is comparatively short, only 1500 h, compared with 5000 h for the same lamps at 250 V. However, where the working voltage is 110 V, a blended lighting scheme shows certain advantages, mainly in running costs.

250-V system

250 V is probably the most widely used voltage for lighting. At this voltage the disadvantage of the 110-V mercury-discharge lamp (comparatively short life) is overcome. As previously mentioned, the present rated life of a mercury-discharge lamp on 250 V is 5000 h.

From the safety aspect 250 V is obviously not as safe as 110 V, with 55 V to earth, but by again using the principle of double-wound transformers with secondaries, centre-tapped and earthed, the voltage to earth can be restricted to 125 V. However, while 250 V dc is covered by exemption 1 of the Electricity regulations, 250 V ac is not, so that the appropriate regulations must be complied with. Centre-tapping and earthing the midpoint of the transformer secondary does not bring the system within exemption 1. There is no reason why any danger should exist from a 250-V system provided that it is properly installed and maintained. Any possible dangers that it may have are far outweighed by its advantages for use with high-wattage lighting installations.

440-V system

The safety aspect becomes still more important when 440 V is used for lighting. 440 V is the nominal standard power-supply voltage in the UK, and technically there is no reason why this voltage should not be used for lighting. Difficulties can arise with the human element, however, compared with power circuits, lighting tends to be regarded as innocuous. This is probably the result of the 'do-it-yourself' fashion in domestic surroundings. If, therefore, such high voltages are used for lighting, those engaged on maintenance must be alive to the attendant dangers, and be more wary of the practices that are sometimes adopted on lower-voltage systems. This is obviously still more important where two different voltages for lighting are in use on the same works.

From the economic angle, and if the application is suitable, there are substantial advantages to be gained by using the high-output lamps available for use on comparatively high voltages.

Miscellaneous considerations

Any form of lighting must cater for all the requirements that may be called for. One requirement that is often called for is colour discrimination, for example, the ability to distinguish one colour of order card from another. Compared with tungsten-filament lamps, discharge lamps fall down heavily on this count owing to the natural characteristics of the lamp. For example, the mercury-discharge lamps has negligible red in its colour spectrum. To improve the colour rendering of a lighting scheme based on these lamps, the missing colours must be introduced by other means. Two methods of improving the colour of mercury-discharge lighting have now become generally adopted. As already indicated, one is to mix it with tungsten filament lighting, and the second is to coat the inside of the lamp with a red-biased emitting fluorescent powder. This is the principle of the well known fluorescent lamp. This principle is now used widely in high-wattage industrial lamps, the M.B.F. series. By adopting either of these schemes, colour-rendering quantities of the installation can be brought to acceptable standards. There is also a psychological aspect to be considered. Even when no colour discrimination is necessary for the class of work being done, uncorrected mercury-discharge lighting is 'cold' lighting owing to the preponderance of green and blue in its spectrum, and those working in its area look 'sickly'. This is generally considered to be undesirable, and uncorrected mercury-discharge lighting for indoor use is losing popularity for this reason alone.

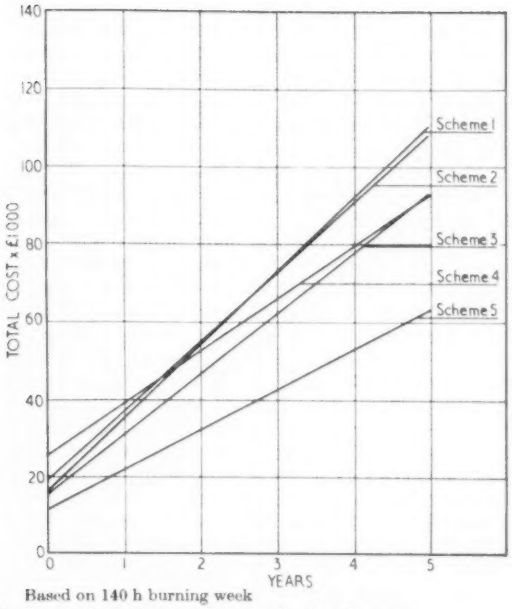
Two disadvantages of any form of discharge lighting should be mentioned. First, there is a stroboscopic effect which gives a flicker effect on moving objects. This has a nuisance value which can be reduced to some extent by suitable methods of connexion.

The second disadvantage is that in the event of a failure of the supply, discharge lamps do not light immediately the supply is restored. The lamps have to cool before the arc will re-strike, and this may take up to about 5 min. This obviously can present certain difficulties, depending on the work being done in the lighted area. However, if a modicum of light is not otherwise available from local lighting or from the process itself, say from furnaces, this disadvantage can be overcome by installing tungsten filament lamps at intervals in the system.

All things considered, there is obviously much to be said for discharge lighting that is colour-corrected either by using it in conjunction with tungsten-filament lighting or by using fluorescent lamps. Based on the cost over the life of the installation, however, the fluorescent lamp is to be preferred, even in spite of the very high initial cost. It should be pointed out that the high first cost is due mainly to the cost of the lamp itself. It is probably a fair assumption that as this type of lamp becomes more generally used, its cost will fall, thus further enhancing the case for its use.

Still comparing the first cost of an installation with the running cost, the curves in Fig.3 show the cumulative cost of the installations under review for the first five years of their life. The comparative steepness of the slope of the curve of the tungsten-filament installation emphasizes the heavy running cost associated with this type of installation.

As a matter of interest, a family of curves has been derived from consideration of capital cost and running cost of a 300-W tungsten-filament lamp and an 80-W fluorescent tube, both lamps having a comparable lumen output.



Based on 140 h burning week

3 *Comparison of cumulative cost for first five years*

The relationship is plotted for a series of burning hours as shown in Fig.4. These curves show that the extra capital cost of a fluorescent fitting can be offset by the lower running cost. For example, if the average burning hours are 40 per week, then for the same total cost over 15 years (life of fitting), it is permissible to spend up to £21 for a fluorescent fitting compared with £2 for a tungsten lamp fitting.

Having emphasized the cost aspect of lighting, some comment is called for on how lighting costs can be kept to a minimum.

Hours of burning

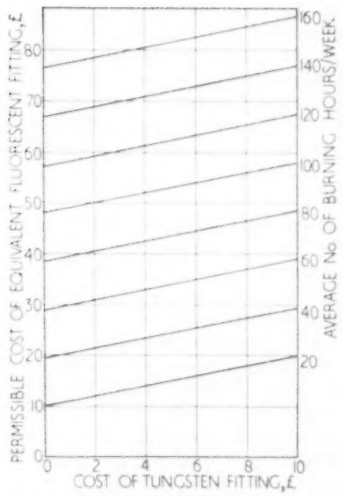
It has already been shown that the most expensive element in the cost of lighting is the cost of power. It follows, therefore, that if the number of burning hours can be reduced, there will be a proportionate reduction in running cost.

There are several well established devices that can assist in this direction, using the principle of the photoelectric cell. For practical reasons, means must be provided to override these devices, and there is a constant problem of the arrangements being left permanently in use, thereby defeating the object of the photocell. A timer used in conjunction with the overriding circuit which ensures that the photocell reassumes control automatically after a preset time can be used to overcome this difficulty.

Lighting intensity

Another important consideration must be to ensure that gross overlighting does not occur. On the face of it, such an observation might appear trite, but there are indications today that such a trend does exist. Very high values of lighting intensity are being used for some steelworks areas which are totally uncalled for. In many cases, half of these values would be more realistic.

It must be added that where 'realistic' lighting values are used, the same realism must be applied to



4 *Curves showing relationship between weekly burning hours and capital cost of fittings of an 80-W fluorescent tube and a 300-W tungsten-filament lamp*

the maintenance of the installation. It is cheaper to design a scheme to an adequate intensity and maintain it well than to overlight as a policy and neglect maintenance. The following example shows this.

Consider an installation designed to 6 lumens/ft² and consuming 150 kVA. Again basing the cost of electrical power on £7 2*s*. per kW maximum demand charge and 1*d*. per unit consumed, and assuming 140 burning hours per week, then:

Total annual power cost = £5615

If the maintenance is such that the efficiency of the installation can be kept to 80%, then

Cost of useful power = £5615 × 0·8
= £4492
Cost of wasted power = £5615 × 0·2
= £1123

A drop in efficiency of 20% means that the value of illumination is reduced from 6 to 4·8 lumens/ft², which is still acceptable.

If the efficiency is allowed to drop to 50% by inferior maintenance, then

Cost of wasted power = £5615 × 0·5
= £2808

At this reduced efficiency, the illumination would fall to 3 lumens/ft², a figure that is certainly too low even for the large working areas under review, and steps should be taken to increase the value. However, before it is necessary to spend capital on additional fittings, etc., £2808–£1123 = £1685 per year could be spent on maintenance to achieve the same object.

Alternatively, if it is decided to accept a low standard of maintenance and let the efficiency of the system fall to 50%, basing the intensity on the low acceptable figure of 4·8 lumens/ft², the installation must be designed to 9·6 lumens/ft². Taking increase in the power requirements *pro rata* with the increase in intensity, total annual cost of power = £8984. An increase of £8984 − £5615 = £3369 per year.

This does not take into consideration the additional capital that must be spent in providing an installation capable of producing 50% more light.

This argument is dependent on a standard of lighting being accepted and maintained. This attitude towards lighting tends to be somewhat diffident. Unlike shortage of cooling water or lubrication, within limits production will carry on with reduced lighting, so that the case for spending money on maintaining the standard of lighting is more difficult to prove.

MAINTENANCE

It has been shown that adequate maintenance arrangements on any lighting scheme are both desirable and economic. There are, however, certain practical considerations that arise on an iron- and steelworks which tend to make maintenance not quite so straightforward as it might appear at first sight.

Because of the variety of the functions that contribute to the production of iron and steel, for form of lighting required varies considerably. The following shows the make-up of the range of fittings used on a lighting installation on a coke-oven plant:

Commercial	171
Open industrial	1049
Open industrial, pole-mounted	78
Flameproof	491
Bulkhead	422
Floodlights	89
Total	2300

Much of the lighting may be classed as local and as such does not fall into any regular pattern in quantity; the 'density' of the installation on this particular plant is only one fitting per 90 yd^2. This results in a high time-consuming factor in maintenance, for the following reasons.

First, consider lamp changing. Assuming an average burning time of 120 h/week (a lower average burning time has been assumed here because of the influence of the comparatively short-time burning of office lights) and an average lamp life of 1150 h (the hours at which 50% have failed, from a typical survival curve), then the replacements required per week.

$$\frac{2300 \times 120}{1170} = 240$$

Observation over a period shows that, owing to the diverse character of the lighting installation, the average time taken to change a lamp (including walking time, dismantling fittings, etc.) is about 10 min. Therefore, the man-hours required per week for lamp changing alone is 40. This figure is for one man only, but in fact in some cases, say for safety reasons, two men are required to work together.

This means that on a plant of this size, and based on a 42-h working week, one man is almost fully employed changing lamps.

The example given assumes that all lamps are tungsten-filament type, but the conclusion will still be fairly correct on plants with mixed lighting, for on such a plant the proportion of tungsten-filament lamps to other types used would normally be high.

There are many theories, some of them well supported by facts and figures, on the most efficient methods of lamp changing, but these tend to be evolved to cover an installation having a large number of similar fittings in a comparatively compact arrangement.

Now consider lamp cleaning. Lamp cleaning would seem to absorb more time per fitting than lamp changing does, but in fact the reverse is the case. The reason for this is that lamps fail at widely scattered points, and the element of travelling time between fittings is high. When fittings are being cleaned, however, this work can be planned on an area basis, and so the non-effective working time per fitting is reduced. An observed average time for cleaning a fitting is about 6 min per fitting. Again, based on a 42-h week and assuming no 'idle' time, 420 fittings could be cleaned every week. On the plant quoted, this would mean all fittings could be cleaned every $2300 \div 420$ weeks, i.e. $5\frac{1}{2}$ weeks approximately. This is a reasonable time cycle. In fact, what would happen in practice is that some lamps would be cleaned more frequently, others less frequently, giving an average cleaning cycle of $5\frac{1}{2}$ weeks.

Unfortunately, iron- and steelworks areas seldom exist in convenient sizes to employ multiples of one man full time on lighting maintenance and, in most cases, one man may be doing lamp changing and lamp cleaning. This reduces the efficiency of working still further for, while lamps fail in different areas, they also fail at varying hours of the day. This means that if burnt-out lamps are to be replaced soon after they fail (and in some cases this may be necessary), the lamp-cleaning programme is interrupted, the non-effective working time increases, and thus the efficiency of the cleaning operation falls. For this reason alone it would seem that lamp changing and lamp cleaning should, if at all possible, be kept as two separate functions.

A further practical difficulty must be mentioned. In most plants, lighting maintenance personnel are semi-skilled as are craftsmen's mates, and occasions arise, due to absenteeism or breakdown conditions, when the need for semiskilled labour is greater on work other than lighting maintenance. Under these circumstances, lighting maintenance inevitably suffers and the work cycle falls behind. It is important, therefore, that this point be catered for in some way when planning a lighting maintenance schedule.

In conclusion it is hoped that the paper has emphasized a few aspects of lighting that are in some way peculiar to the iron and steel industry and present problems which cannot be solved by standard methods in every case. It is felt that surrounding these problems there is scope for much study and thought especially in the field of lighting maintenance. But it is pointless to evolve a system which works only in theory. The proof of the efficiency of the system is that it is a long-term practical and economic success.

Further, it has been shown that good lighting is an expensive commodity, though a necessary one. However, with a realistic approach to design, for example, by lighting for requirements and not for show, by long-term rather than short-term costing, at least the expenditure involved can be justified. It should, perhaps, be stated at this point that the curves produced are based on certain general assumptions. While these assumptions may be varied to suit individual cases, it will be found that the trend of the curves is the same in each case.

Finally, the importance of good maintenance cannot be over-emphasized. Good maintenance is the best way to ensure that value for money is being obtained from any lighting installation.

Engineering aspects of the Redbourn development scheme

P. M. Hesketh

SYNOPSIS

The paper deals with the engineering aspects of the Redbourn development scheme which involves major improvements in each of the production departments of an integrated iron- and steelworks. The main objectives of the scheme are outlined, an account is given of the various engineering developments, and some problems which have arisen are described in detail. 1892

INTRODUCTION

ALTHOUGH the Redbourn scheme has not yet reached completion, the extent of the developments and the circumstances under which they are being carried out led to the suggestion that even at the present stage a paper would be of general interest. The subject has been dealt with in three sections: (*a*) a brief outline of the scheme and its objectives, (*b*) a detailed account of the engineering developments, and (*c*) special problems which have arisen.

OUTLINE OF THE DEVELOPMENT SCHEME

The main objectives of the scheme were first, to raise steel output from less than 10000 ingot tons to at least 14000 ingot tons/week, and second, to increase the quantity of billets rolled from about 6000 tons to about 12000 tons/week. At the present time about 4000 tons/week of slabs are sent to Ebbw Vale and no mill capacity is available to roll this material into billets. When the developments have been completed, it is intended that the full output of steel should be rolled to billets.

To achieve these objectives it was necessary to carry out the following developments:

Mines
The development of a new open-cast iron-ore mine at Winterton which will increase the reserves of local ironstone by some 25 million tons.

Coke ovens
The rebuilding of 41 old coke ovens and the construction of a new battery of 62 ovens, to raise the output of coke from some 8000 tons/week to about 10500 tons/week.

Ironmaking
An increase in ironmaking capacity, first by enlarging the diameters of the existing furnaces, and secondly by the installation of a modern ore preparation and sinter plant which will feed to the furnaces some 24000 tons/week of high-grade sinter. Together, these developments will raise the iron make from about 7000 tons to about 11000 tons/week, and at the same time will give a lower coke rate for a better quality iron.

Steelmaking
The installation of the rotor steelmaking process to pre-refine hot metal before delivery to the open-hearth furnaces. This treatment will reduce the tap-to-tap time from about 11 h to about 8 h.

Mill
To raise the billet rolling capacity from 6000 to more than 12000 finished tons, the old reversing mill is being replaced by a continuous billet mill comprising four roughing and four finishing stands. By the addition of a further two roughing stands this mill will be capable of dealing with up to 20000 ingot tons should the need arise.

Engineering shops
To accommodate the new continuous billet mill adjacent to the existing cogging mill and soaking pits, it has been necessary to transfer the engineering workshops to an entirely new building. The developments will also demand better maintenance facilities and these are being provided in these new central engineering workshops.

The programme on which this scheme is being carried out is shown on the bar line chart in Fig.1. Bearing in mind that these major construction programmes are being carried out concurrently while the plant is operating at normal levels of production, it will be appreciated that the engineering planning must be of the highest quality.

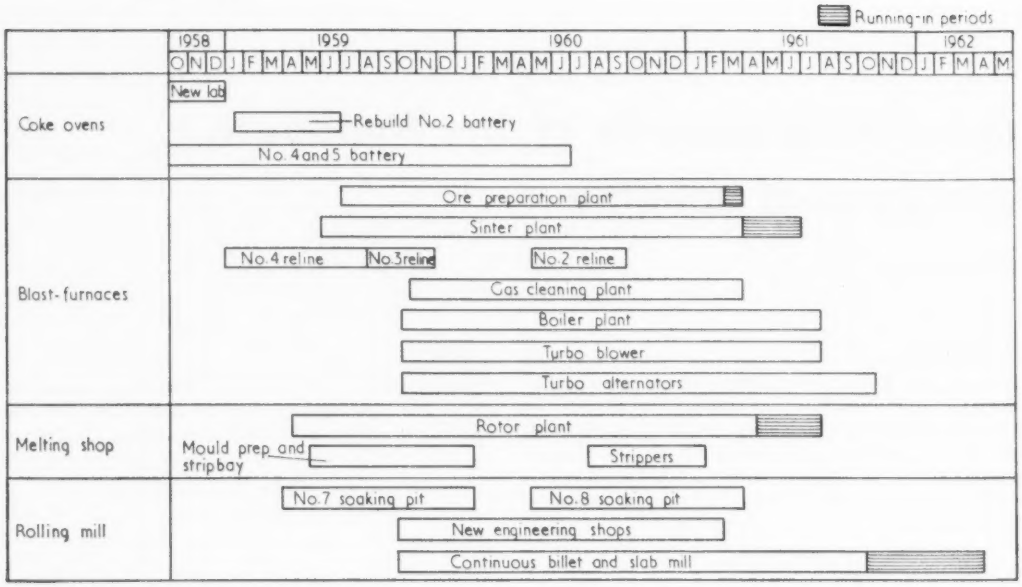
DETAILED ACCOUNT OF THE ENGINEERING DEVELOPMENTS

Coke ovens

Before development, the coke-oven department comprised two batteries of 41 ovens which had been in operation since 1938 and a 19-oven battery which was completed in June 1955.

The programme was to build a new 62-oven plant (two batteries with 31 ovens each) and to rebuild completely one of the 41-oven batteries. When the new 62-oven battery is commissioned the other 41-oven battery will be demolished and the 'pad' retained for future development. The by-product plant is being increased in capacity and provision is made in the

Manuscript received 22 August 1960.
The author is chief engineer, Redbourn Works, Richard Thomas and Baldwins Ltd.



1 *Development programme at Redbourn*

scheme for a gas booster house and pipelines to export coke-oven gas to the gas grid.

A major problem which arose was that of feeding coal to the 19-oven battery during the reconstruction period, the 41-oven battery being between the 19-battery and the service bunker. Coal had to be discharged from the bunker on to a cross conveyor then down a chute to a longitudinal conveyor running the length of the 41-battery on the pusher side at ground level. It was then elevated by means of a vertical and horizontal Redler conveyor and fed directly into the larry car serving the 19-oven battery. The longitudinal conveyor had to be bridged over to allow access to the rebuild operation. Fortunately this system worked extremely well and gave little trouble.

Ore preparation and sinter

The present practice is to crush the ore down to 5 in. and charge this stone, including the fines, directly on the furnaces. Under the development programme, a new ore preparation and sinter plant is being constructed. This will comprise primary, secondary, and tertiary crushing, the local stone passing directly from the secondary to tertiary crushing, but the Northants stone passing via ore dryers after secondary crushing to tertiary crushing.

The plant will be capable of a throughput of 500 tons/h but the main crushers and conveyors will be capable of 800 tons/h. Coke rod mills are being provided together with facilities for adding flue dust into the system. The sinter plant comprises two separate strands 8 ft wide and 208 ft long, with a rated output of 16000 tons/week per strand. Provision has been made for the installation of a third strand.

Blast-furnaces

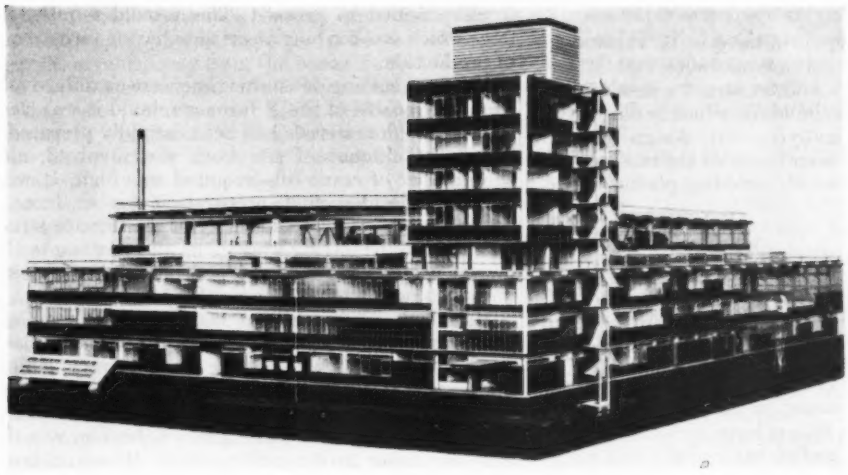
The main units are three bucket-charged blast-furnaces with their ancillary plant. The major items in the programme were as follows: No.4 furnace was increased from 17 ft 9 in. dia. to 19 ft hearth dia. The 'horizontal–vertical charger' was replaced by an inclined hoist, still retaining the bucket-charging principle, similar to that in use on the other two furnaces, and three new hot-blast stoves were constructed. No.3 furnace was also increased from 17 ft 9 in. to 19 ft dia. No.2 furnace has been increased in diameter from 19 ft to 22 ft.

Blast-furnace gas-cleaning plant

To cater for the increased blast-furnace gas make, the existing gas-cleaning plant, consisting of five scrubber towers and five electrofilter units, is being extended by a further scrubber tower and electrofilter unit. This will increase the cleaning capacity by $2\frac{1}{2}$ m. ft^3/h. It is normal during three-furnace operation to have one unit off for cleaning and overhaul. Provision has been made for the addition of a seventh unit.

Water from the gas-cleaning plant is treated in a Dorr thickener and the underflow passed to a disc filter unit to enable the sludge to be more easily transported. The Dorr thickener was inadequate for dealing with the water after development and increased capacity had to be provided. Originally the scheme was to provide an additional thickener but this would have taken a considerable amount of space and would have been rather costly. Finally, it was decided to increase the depth of the pond by about 10 ft and to change the centre mechanism to a Clari-Flocculator. This has increased the capacity from a nominal 240000 gal/h to 416000 gal/h and has given an overflow containing not greater than 3–4 grains/gal. Work on the civil engineering construction to increase the depth of the pond was begun in January 1960 and completed in July 1960. During this period the Dorr thickener remained in operation.

The thickener was taken out of commission during the annual stop period, two blast-furnaces only being in operation, the water from the gas-cleaning plant



2 *Model of central engineering shops*

being passed through a temporary settling pond. The centre mechanism was removed and a new Clari-Flocculator mechanism built up together with new flumes, etc., the unit being returned to service after 10 days.

Turbo blowers

The existing turbo blowers are:

Two BTH, each rated at 60000 ft^3/min at 15 lb/in^2 gauge

One Escher Wyss rated at 40000 ft^3/min at 17·25 lb/in^2 gauge

One Fraser and Chalmers rated at 40000 ft^3/min at 15 lb/in^2 gauge.

The last unit operates with low-pressure steam and is kept as a standby.

None of these blowers will be capable of supplying the necessary volume of air to the enlarged no.2 furnace when operating on sinter. Therefore it was decided to install a Richardson Westgarth axial blower to cater for this demand, and this unit will replace the existing Fraser and Chalmers unit. It will be rated at 55000 ft^3/min at 18 lb/in^2 but by modifying the blading this can be increased to 72500 ft^3/min at 25 lb/in^2.

Melting shop

There are seven 120-ton fixed OH furnaces in the melting shop together with a 500-ton mixer, the iron from the blast-furnaces normally passing to the desiliconizing plant before going to the OH furnaces. It was decided to install a rotor furnace to pre-refine the iron from the blast-furnaces before charging to the OH furnaces. This will reduce the load on the OH and enable the shop output to be increased.

The original rotor vessel at Oberhausen in Germany is capable of rotating about its longitudinal axis, with a small degree of tilt, but the unit to be installed at Redbourn rotates about its longitudinal axis and in addition can be rotated in a horizontal plane. This will allow the vessel to be moved into various positions for charging, oxygen blowing, slagging, tapping, etc., and to be changed end for end to equalize lining wear. The vessel will be tilted for slagging and tapping and will tilt so that the axis of the vessel is vertical. In this latter position the vessel will be taken from the turntable cradle and be moved by winching to a relining position. A spare vessel will then be placed in the turntable and operations continued while relining takes place on the first vessel. The principal details of the plant are:

Capacity of each rotor vessel	85–100 tons
Inside diameter of vessel	14 ft 1 in. approx.
Length of vessel	49 ft 3 in. approx.
Speed of rotation	0·5–5·0 rev/min

During the pre-refining, slag is produced with a saleable P_2O_5 content and a plant for processing this slag is being constructed in the works.

In the early stages of development, four of the OH furnaces have been rebuilt to a new design, and these rebuilds are described later in this paper.

Rolling mill

The main development in the rolling mill is the replacement of the two-high steam-driven reversing billet mill by an electrically driven continuous billet mill. The new mill consists of a four-stand roughing mill and a four-stand finishing mill, the stands being alternately horizontal and vertical and each being independently driven.

This new mill will be served by the existing cogging mill and will be situated in an adjacent bay at present housing the engineering shops. The layout has been designed so that commissioning can be completed while the existing reversing mill is still in production; the old mill will be taken out of service as soon as the new mill has achieved the required performance. An important feature of this new mill is that sizes can be changed and the change made from billets to small slabs in a matter of minutes. A further feature of special interest is the installation of a continuous inspection facility and this is dealt with in some detail later.

As part of the development programme in the mill, the existing cogging mill will be modernized, the main items being new stands, manipulators, roll-changing rig, and scale flushing. Also one new four-cell soaking pit of the 'horseshoe'-fired type with metallic recuperators was commissioned in January 1960, and a similar unit to replace two old pits will be commissioned in the spring of 1961.

Central engineering shops

As the present workshops will have to be vacated to accommodate the new continuous billet mill, it has been necessary to design and construct a new central engineering facility. This building, which is illustrated in Fig.2, incorporates many modern design features and full advantage has been taken of the most up-to-date materials of construction including plastic-coated sheet.

Steam and power

The present steam-raising capacity consisting of three 300 lb/in^2 water tube boilers and 20 Lancashire boilers at 150 lb/in^2 will be increased by the addition of three boilers rated at 100000 lb/h MCR at 650 lb/in^2. The Lancashire boilers will be demolished when the new plant has been commissioned in the summer of 1961. In addition, waste heat boilers have been installed on two of the OH furnaces and steam-cooled lintels are to be installed on one of the OH furnaces.

The steam from the new boilers will pass through a 5268 kW back-pressure turbo alternator, exhausting at a pressure of 300 lb/in^2 for use on the turbo blowers and for supplying the rest of the works where the pressure is 150 lb/in^2. The existing three turbo alternators in the power house comprising two at 1500 kW and one at 3000 kW will be supplemented by a 1958-kW medium-pressure back-pressure turbo alternator exhausting at 155 lb/in^2.

Finally, it should be mentioned that perhaps the most far-reaching changes called for by these developments have been concerned with material-handling facilities. The alterations and extensions to the rail and road layouts within the works have been extensive.

SPECIAL PROBLEMS WHICH HAVE ARISEN

All schemes of this nature must be engineered so as to overcome the obstacles imposed by limitations in time, space, and finance. It would be quite impracticable to outline all the special problems which have arisen within the scope of the present paper. Nevertheless, it is possible to outline the main features of two of these problems and it is hoped that others can be considered when this paper is discussed.

Furnace rebuilds

During the developments, four OH furnaces were to be rebuilt to a new design based on single uptakes, enlarged slag pockets, and the use of retractable fuel burners. This design incorporates a complete basic suspended roof, suspended slag pocket arches, and slag pocket main walls. These changes had already been tested on a previous furnace and had shown improvements on production from that of the original furnace. The rebuild of the furnace involved the complete wrecking down to foundation level before erection of the new structure.

The first to be rebuilt was *A* furnace and all the materials were prepared in our own workshops. Dismantling began on 12 July 1958 and the furnace was completed and gassed on 21 August 1958. The annual 'stop' fortnight occurred during this period and the furnace could not be fully manned as personnel were engaged on other work. This had been realized, but as the furnace would not be operating during this period it was decided to proceed. This rebuild required 40 days which was too long a period to have a furnace out of production.

Means of cutting down this time were examined and from the records of the *A* furnace rebuild it was clear that although materials had been carefully prepared a significant amount of site work was involved; also the number of crane lifts required was high. It was, therefore decided that preparation for *C* furnace would aim at complete assembly of the furnace structure in the engineering workshops, dismantling in the largest possible transportable sections, and re-erection on site using the minimum number of crane lifts.

Had it been possible to transport the complete structure, which was in the region of 350 tons, in say three sections, it would have been a very simple matter to make the fabrication sound enough to travel. Bulk and weight made this out of the question in the engineering shops and crane girder headroom was the deciding factor in the melting shop. However, large sections such as the port ends and block ends were left assembled and this meant the saving of 60 lifts on site. In addition, the building of all foundation sections at the end of the furnace stage further reduced the actual number of lifts during erection. To reduce further the load on the stage crane and to expedite the dismantling of the furnace, a special track and bogie were prepared to be fixed across the furnace stage after gas was taken off the furnace. This track gave access to the adjacent magnet crane bay.

Personnel worked on shifts during the whole of the rebuild. The weight of steelwork erected was over 350 tons and the number of bricks used was 257000 pieces with a total weight of 1200 tons. Despite this, the furnace returned to production only 18 days after wrecking began. Detailed costs taken during the rebuild, including the prefabrication in the workshops, showed that in addition to saving time the method used was cheaper and valuable furnace hours were gained.

For the next furnace rebuild on *B* furnace the same methods were used and, in addition, the furnace pipework for water cooling was prefabricated to reduce further the site work required. This rebuild was completed and flares lit on the 16th day after demolition began. A large amount of extra work was involved in this case as the chequers were in bad condition and it was decided to redesign to give a single chamber at each end of the furnace, using a flat fully suspended roof. During the rebuild, this involved an additional 50000 bricks as compared with *C* furnace.

The fourth furnace has already been assembled in the engineering shops and site work will take place at the end of 1960, and it is hoped to equal or better the previous times. This work on the OH rebuilds has once again shown the value of careful planning and the advantages of doing as much as possible away from the producing unit.

Continuous inspection

Under existing arrangements, inspectors walk on the cooling banks, the bars being turned so that all sides can be seen. Good bars, normally 60% or more of the cast, are sorted and loaded for despatch, defective bars being sent to the dressing area. Personnel include an inspector, two helpers, and a loader. On the new billet mill it became apparent that space available for

cooling banks would be limited and the present method would, therefore, have to be modified and inspection done away from the banks.

It was suggested that if the bars were run off the cooling banks on to a roller table, the bars could pass inspectors who could bypass defects and send good bars direct into a cradle for despatch. This would involve the crane being used only once for the good bars. Tests were made and it was found that inspection could be comfortably carried out with bars passing the inspectors at 200 ft/min. Relating this to the output of various sizes of billets and slabs showed that one inspection line would be sufficient for most sizes and that two lines would cover all outputs.

The first tentative layout gave an inspection line in which the bars first passed two inspectors, one on either side of the rack. The bar was then turned and passed two further inspectors so that all four sides were examined. To cater for 30-ft billets, this gave a long inspection layout. A suggestion was then made that if on leaving the cooling banks the bars were turned on to the diagonal and run on V-rollers, all four sides could be inspected by passing two inspectors instead of four. Tests were carried out and showed that this was quite feasible and reduced the length of the inspection line by nearly 50%. Lighting tests have also been carried out to ensure adequate illumination without glare for two inspectors sitting opposite to one another.

The method of operating this inspection equipment will be to push bars from the four-section cooling bank by means of skids on to a runout table with individually driven rollers running at a speed of 300 ft/min. Groups of bars will then be conveyed into the reach of a chain skid, which has a conveying speed of 100 ft/min, to bring the bars at the option of the operator on to one of two cross-transfers, depending on which of the two inspection lines is being used. The cross-transfer has a skid speed which is infinitely variable within a range of 1–10 ft/min. The bars will move towards the V-roller table at a speed corresponding to the bar section and length. One bar after another will fall at equal time intervals over an inclined plane on to the V-roller table. An automatic mechanism to prevent damage to the V-rollers from falling bars has been designed.

The V-roller tables which have a speed range of 100–300 ft/min will take the bars past the two inspectors. The V-roller table is followed by a further table of cylindrical rollers which has a constant speed of 300 ft/min. In this roller table are arranged three chain skids and a number of disappearing stoppers. The chain skids can be moved either to one side or the other as may be required. When either of the inspectors finds a defect he will operate a push-button which automatically reverses the pushing direction of the chain skids to send the bar into the defects cradle. The good bars go automatically into the necessary cradle. A number of cradles are provided in each line of both good and defect bars, and one of the inspectors can, by selection of the disappearing stoppers, decide which of the cradles is to be used. It is not intended that the defect bars be marked in any way as a single defect will be sufficient to reject a bar, and the inspector need not look for any more faults. It will then be the responsibility of the dressing area to examine all the defect bars and to dress where required.

Although the inspection equipment will not go into operation until the end of 1961, most of the features have been tested and it should provide a quicker method of inspection with reduced crane handling and possibly reduced manpower.

ACKNOWLEDGMENT

The author and his colleagues would like to acknowledge the unstinted co-operation and help which they have received in planning these developments, from operators and engineers in many UK and European companies.

Developments at Appleby-Frodingham since 1954

E. A. Atkin and E. F. Farrington

SYNOPSIS

The paper gives a brief history of Appleby-Frodingham and current plant performances. It then describes developments since 1954 and outlines the recently announced development plans. 1896

INTRODUCTION

IN COMPANY with most other iron and steel plants, the recent history of Appleby-Frodingham has been one of continued expansion and improvement of its production processes.

The work done since 1954 can only be briefly described, but it is hoped that sufficient has been included to make the paper of general interest and that such particular omissions as appear to individual readers will only serve to stimulate discussion at its presentation.

Manuscript received 24 August 1960.
The authors are with the Appleby-Frodingham Steel Company.

HISTORICAL BACKGROUND

The plant is the result of the amalgamation of two separate companies, the acquisition of a third, and the

TABLE I Appleby-Frodingham Steel Company production data 1954–59

Product	53–54	54–55	55–56	56–57	57–58	58–59
	Tons	Tons	Tons	Tons	Tons	Tons
Ore crushed	3556131	4324441	4103092	4691535	4506267	4940963
Coke made	655404	648175	660516	657331	612399	664365
Sinter	1958820	2977770	2843750	3494750	3347680	3800240
Iron	978270	1201070	1127980	1342730	1292470	1389650
Ingots	1059701	1029810	1060446	1172105	1176595	1126846
Sections	341286	346591	357324	383175	374114	289843
Plates	420797	395912	435056	450460	442341	450488
Basic slag	124563	104870	121877	120155	121973	139134

extension of its territories over contiguous areas following ore mining.

The Frodingham Iron Company began operations in 1865, and the Appleby Iron Company in 1874. In 1888, the Frodingham Company started steelmaking and rolling, and in 1902 installed the first open-hearth tilting furnace in Europe.

In 1912, the Frodingham Iron and Steel Company, as it had been since 1909, acquired the Appleby Iron Company, and a development plan was prepared which provided for the Appleby company installing two plate mills. Owing to adverse trade conditions, among other things, these plans did not mature until 1927. In the meantime, in 1918, both the Frodingham and the Appleby companies had joined the newly formed combine, The United Steel Companies Ltd. In 1934, the Frodingham and Appleby companies combined to form the Appleby-Frodingham Steel Co. Ltd. Finally in 1946, the Appleby-Frodingham Steel Company became a branch of The United Steel Companies Ltd.

PRESENT PERFORMANCES

Today, Appleby-Frodingham directly employs 11700 people. Traditionally, it produces plates and sections, based upon the mass production of mainly ordinary quality steels in basic OH tilting furnaces from phosphoric pig iron derived from home ores.

Present capacity is 1·6 m. tons of pig iron and 1·2 m. tons of ingots per year; additional production data is given in Table I.

The plant occupies 1300 acres, is served by 110 miles of internal railways, 47 locomotives, and 12 miles of roadways. It receives nearly 8 m. tons/year of raw and other materials, despatches about 1½ m. tons/year of finished and other materials, and handles 7 m. tons/year of internal traffic exchanges. It uses 375 m. units of electricity and 6 m. tons of water per year.

PLANT DEVELOPMENT FROM 1954 TO 1960

The developments are described under the broad classifications, coke ovens, ore preparation plant, blast-furnaces, melting shops, and finally rolling mills. The works plan (Fig.1) shows major developments completed since 1954.

Coke ovens

The original installation of 66 ovens was extended in 1951–52 by the addition of nos.3 and 4 batteries, thus doubling plant capacity. The use of blast-furnace gas for underfiring was leading to the fouling of the regenerators on nos.1 and 2 batteries. To prolong their life as well as to avoid similar trouble occurring on the rest of the plant, a horizontal electrostatic precipitator was built to clean the blast-furnace gas supply to the coke ovens. This was completed in 1958.

In 1955, to secure additional coke of the right quality for the blast-furnaces, nos.5 and 6 batteries were ordered with the object of working a five-battery plant (i.e. five working and one rebuilding). The resulting 25% increased capacity led to the posing and solution of some complicated problems on the chemical recovery plant. These batteries were completed in 1959, but owing to trade conditions then existing, they were not lit until the end of November 1959. All six batteries are now working at a level of production equivalent to standard five-battery operation.

A new pure-products plant has just been completed to replace an earlier plant built for the first two batteries. Work is nearing completion on the construction of a plant to desulphurize 32 m. ft^3 of coke-oven gas per day.

Ore-preparation plants

In 1955, an increase in ironmake was authorized. Since commissioning in mid 1954, the Seraphim sinter plant had shown itself to have a greater potential capacity than that for which it was designed. The increased ironmake required was achieved by adding units to the Seraphim ore preparation plant, and by modifying the sinter cooling and return fines system on the Apex plant to bring it into line with proved features on the Seraphim plant.

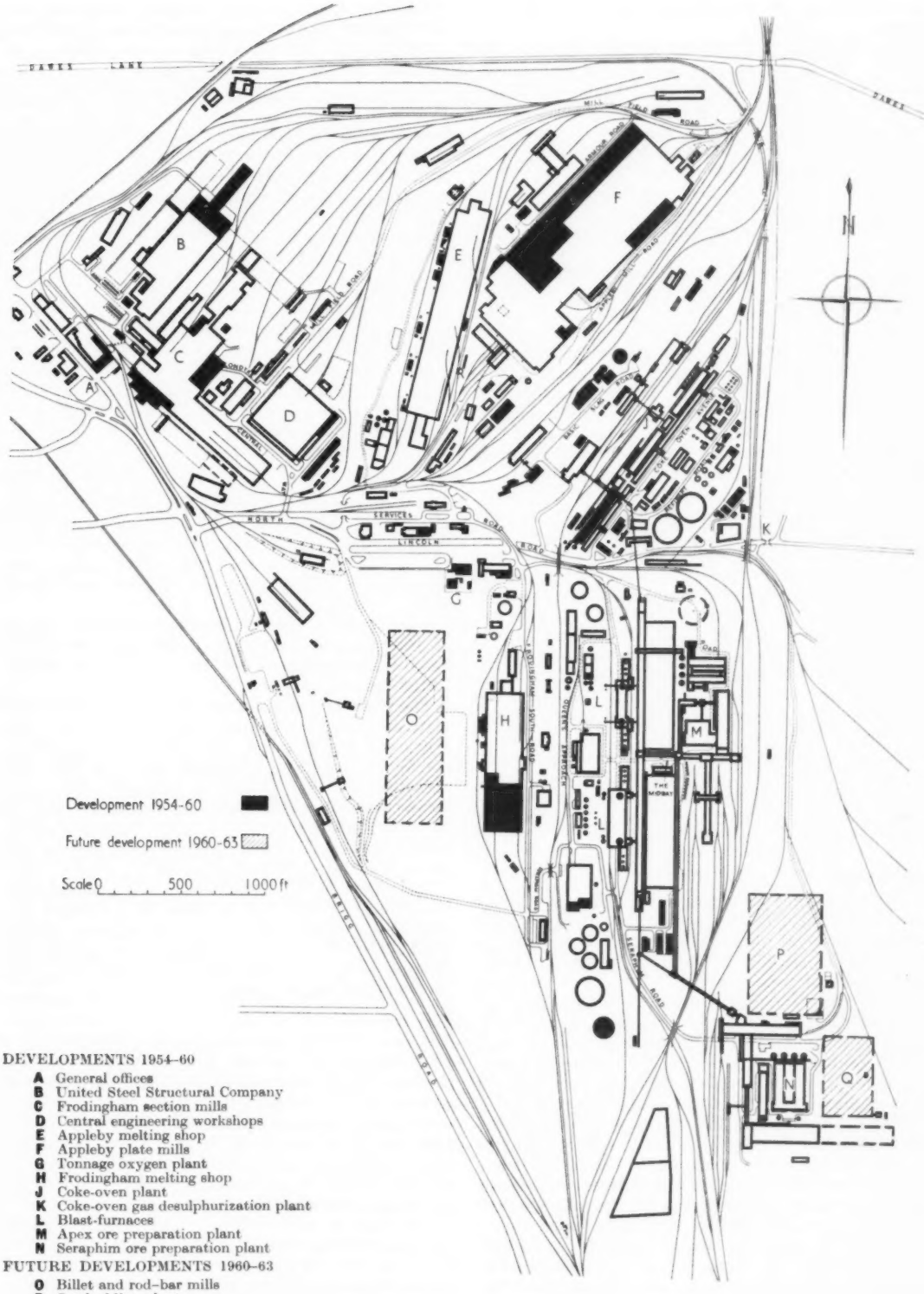
The work done on the Apex plant deserves mention. The natural draught sinter coolers had never worked satisfactorily and it had become necessary to spray the finished sinter with water to prevent burning the finished sinter conveyor belts. The quenching water caused intolerable working conditions and a high percentage of fines (24% < ¼ in. on the Apex plant compared with 10% on the Seraphim).

Similarly, the Apex return fines system, judged by the improved standards then required, was inadequate to effect the proper degree of control over the raw mix and return fines additions. A new system was installed to extract the return fines to a separate plant where they could be cooled, screened into two sizes, and fed back into the raw feed systems at a controlled rate. This work was completed late in 1956. In 1959, an experimental granulating drum was introduced to feed one of the Seraphim sinter strands.

Blast-furnaces

In 1954, the four-furnace plant had an initial capacity of 23500 tons/week. Furnace trials, using Seraphim sinter, revealed that it had a potential of 29000 tons/week given additional blast air.

Thus when the additional ironmake was authorized in 1955, it was only necessary to carry out the work already described on the ore preparation plants, and



DEVELOPMENTS 1954–60

- **A** General offices
- **B** United Steel Structural Company
- **C** Frodingham section mills
- **D** Central engineering workshops
- **E** Appleby melting shop
- **F** Appleby plate mills
- **G** Tonnage oxygen plant
- **H** Frodingham melting shop
- **J** Coke-oven plant
- **K** Coke-oven gas desulphurization plant
- **L** Blast-furnaces
- **M** Apex ore preparation plant
- **N** Seraphim ore preparation plant

FUTURE DEVELOPMENTS 1960–63

- **O** Billet and rod–bar mills
- **P** Ore bedding plant
- **Q** Sinter machines
- **R** Continuous casting

1 *Works plan*

to add blowing and gas-cleaning capacity to the ironworks.

The blowing plant added consisted of a 150000 ft³/min, 35 lb/in² 'split wind' centrifugal blower driven by a 20000 hp steam turbine supplied from a blast-furnace gas (alternatively oil) fired water tube boiler rated at 135000 lb/h at 450 lb/in² gauge and 760°F. This entailed alterations and additions to the condenser and gas-cleaning plant water supply and cooling arrangements. In the original Seraphim installation water for cooling, arising from blast-furnaces, gas cleaning, and condensers, was handled in a large three-zone cooling tower. Experience had shown that gas-cleaning water was contaminating the other two owing to droplet exchange in the shaft of the tower. When the new blowing plant was added, more cooling capacity was required and this was provided by building a new tower for gas-cleaning plant water, thus making the original gas-cleaning section of the large tower available for condensers. The additional blower allowed the reciprocating gas blowing engines supplied in 1939 for the first two furnaces to be scrapped. An additional wet electrostatic precipitator, making six in all, was added to the Seraphim gas-cleaning plant.

The increased blowing rate coupled with near 100% sinter burdens gave rise to failures on numerous parts of the blast-furnace plant, e.g. bells and hoppers, hot-blast stove valves and seats, and bleeder valves and seats. In March 1958 the large bell and hopper in one of the smaller furnaces failed and had to be changed. The operation was carried out in seven days with the furnace in the 'damped down' condition. In the following September, a similar operation was carried out on the second small furnace. Attention to design and materials have now substantially overcome the major problems of engineering maintenance on hard-driven blast-furnaces.

Since that time, further improvements in techniques supported by further furnace trials, have allowed iron production to be raised to around 31000 tons/week and have paved the way for the next major advance.

Melting shops

In 1948, following the construction of the new Frodingham melting shop, the production capacity of steel melting plant was about 800000 ingot tons/year. By 1956, without significant additions to capital equipment, this had been raised to just over 1 m. ingot tons/year. This increase was due to the inception of continuous working on the melting shops, to improved operating techniques and improved materials, as well as to a gradual improvement in the quality and consistency of the iron supplied from the blast-furnace plant as the result of the developing sinter practice.

In 1956, an additional 350-tons capacity OH furnace was built in the Frodingham shop to supply ingots to another branch of the combine. At the same time a number of additional service installations, such as an ingot stock gantry, a refractory grinding plant, a new basic slag plant, and steel foundry, were built.

In 1956–57 attention was directed to ways and means of increasing output from the existing steelmaking plant. This led to the establishment of the Ajax process and to the conversion of the first OH furnace (*B* furnace in the Frodingham shop) in January 1958. Since then two further furnaces have been converted (*L* and *N* in the Appleby shop) and more are planned. This work has already been described by Jackson *et al.** To supplement oxygen supply, a 100-tons/day oxygen plant is being built.

Plate mills

Here a major programme of re-equipment is in its final stages, the object of which is to raise production and to improve the quality of the finished plate. In 1954, the total finished production was about 400000 tons/year. Now, in 1960, it is about 500000 tons/year. The layout of these mills is shown in Fig.2 (the re-equipment scheme is shown shaded).

Soaking pits

Since 1954, two four-cell one-way fired recuperative soaking pits have been installed, one in place of an original two-way fired regenerative type. One electric soaking pit was added in 1957, and one controlled cooling pit in 1959. In preparation for the above, two new soaker cranes were installed in 1954–55 which required new crane girders and the raising of the roof over the building.

7-ft shearline

A new shearline was commissioned in 1957. It consists of a rotary trimmer–slitter shear with associated plate-handling devices and scrap disposal mechanisms, and two sets of cross-cut shears with length gauges, scrap shears, and piling tables. A new despatch bay was also built.

The line is designed to have a throughput capacity of 4000 tons of finished plate per week of 127½ h, based on an average plate of 59 in. wide × ¼ in. thick × 34 ft long.

The trimmer has two heads capable of admitting a plate up to 84 in. wide, one fixed and one sliding on a heavy cast-iron bedplate. The rotary blades are 31½ in. dia. and are equipped with means of vertical and axial adjustment. The trimmer is equipped with scrap choppers capable of dealing with 6-in. wide scrap up to ⅜ in. thickness, and 4-in. wide scrap up to ¾ in. thickness. The choppers discharge on to a scrap conveyor, and thence into scrap boxes.

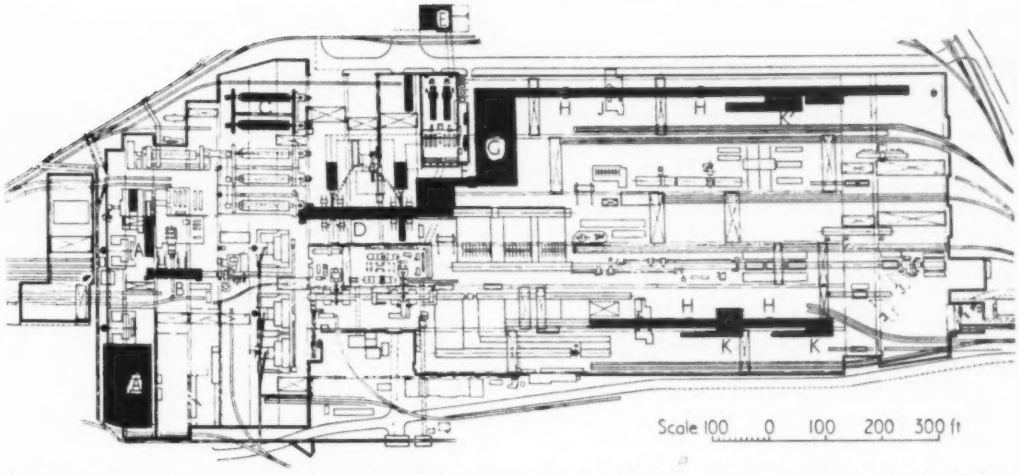
The slitter head, which is equipped with the same blades as the trimmer, is mounted on a separate bedplate and is capable of slitting to a minimum width of 20 in. The two cross-cut shears were rebuilt from existing equipment and each is provided with scrap choppers and length gauges.

Slabbing mill

The new slabbing mill was completed in December 1959. Because of defects in one of the main housings, the installation had to be carried out as two separate operations. It had been the intention to install the new mill in the annual two-week shutdown in July of 1959, but in fact only the roller tables and manipulator equipment could be installed then, and the housings had to be installed later during a special six-day stoppage at Christmas 1959.

The mill is a conventional two-high, high-lift slabbing mill having rolls of 114-in. barrel length and 45 in. dia. The rolls are hydraulically balanced by an independent accumulator system and are supported in

* A. Jackson *et al.*: *JISI*, 1958, **190**, 1–29.
* A. Jackson: *ibid.*, 1959, **191**, 337–342.



A Soakers **B** Slabbing mill **C** Slab reheaters **D** 12-ft plate mill **E** Electrical equipment **F** Hot mangle **G** Cooling bank **H** Cross-cut shears **J** Slitting shears **K** Piling tables

2 *Appleby plate mills*

fabric bearings. The roll assembly is supported on a retractable sledge to facilitate quick roll changing.

The screwdown gear, powered by two 150–300 hp motors designed for speeds up to 220 in./min, is equipped with automatic screw control initiated by photocells on each side of the mill. The breaker block on each screw is fitted with load cells, and the reading is transmitted to the control pulpit. Provision is made for cancelling the automatic screw control, and the operator can switch from automatic to hand control instantaneously. The mill is operated by one man.

12-ft plate mill

The original plate mill, built in 1926, consisted of a 12-ft roughing and a 10-ft finishing mill, both two-high, built in line on a common drive. This system had worked well for 30 years, but could no longer meet the current demands for quality and output, and maintenance costs were rising.

It was accordingly decided to install a new 12-ft four-high finishing mill downstream from the original, and a new 12-ft two-high roughing mill in tandem with it. To cater for the expected increased production, it was also decided to install a second plate-shearing and finishing train for light wide plates, to supplement the original which was only 10 ft wide and which could not conveniently be made wider. This work was carried out in three phases.

The first phase was to replace the 10-ft finishing stand with a 12-ft wide, four-high unit driven by twin motors. The second was to supplement the existing shear line by a complete new rotary shearing and conveying line for 12-ft wide plates; and the third was to replace the 12-ft roughing stand by a new twin-motor driven two-high stand.

The four-high finishing stand was installed during the two-weeks' annual shut-down in August 1958, the shearline during 1959, and the two-high roughing mill in July 1960.

Two additional bogie-type reheating furnaces (making five in all) having 1½ times the heating capacity of the original were built as part of this development.

12-ft finishing stand

On the finishing stand, the back-up rolls are nominally 63 in. dia. dressing down to a minimum of 59 in. dia. and have a barrel length of 150 in. The rolls are mounted in Morgoil bearings. The work rolls, nominally 39 in. dia. dressing down to 37¼ in. dia., are 150 in. long on the barrel and are mounted in Timken four-row roller bearings. The roll assembly is mounted on a sledge and is balanced hydraulically by a separate weight-loaded accumulator. The work rolls are driven by 35-ft long spindles directly coupled to two 4000-hp 40–100 rev/min motors.

The main table rollers, spaced at 48-in. centres, are of solid forged steel 18 in. dia. mounted in self-aligning roller bearings, the bearings being housed in hardened steel sleeves. The rollers are driven by lineshafts having 'Bratt'-mounted machine-cut bevels. The whole lineshaft is adjustable by means of a worm-wheel assembly which also acts as the thrust assembly.

Each side of the mill is equipped with heavy side-guard manipulators having a minimum opening of 3 ft and a maximum of 13 ft. The manipulators are electrically interlocked so that the delivery side opening is always 7 in. wider than the entry side. On the return pass the position is reversed.

A television camera is set up on the outgoing side of the mill and relayed to the roller and the driver. The roll load is measured and indicated to the roller by load cells. The mill is controlled by one man.

The mill is serviced by a 120-ton crane. Work roll changing is carried out by a porter bar, this operation being carried out during the ½-h meal break. The back-up roll change takes about 12 h and is carried out at weekends.

12-ft side trimming and slitting line

The line is designed to accept full width and length plates from the finishing mill in thicknesses between $\frac{3}{16}$ and $\frac{3}{4}$ in. Plates above this thickness are routed through the heavy shearline which is only 10 ft wide.

The $\frac{1}{4}$–$\frac{3}{4}$-in. thick plates leaving the finishing stand are conveyed by an overhead pusher across a transfer bank to the ingoing table of a hot leveller. This has nine levelling rolls supported by back-up rolls and is

capable of levelling 146-in. wide plates of between $\frac{3}{16}$ and 1 in. thickness.

Plates leaving the leveller are transferred on to a cooling conveyor bank, the conveying being performed by fingers housed in moving heads. The heads are actuated by a shaft linkage system, all plates on the bank being moved progressively forward.

The plates, at the end of the cooling bank traverse, arrive at a turnover gear for inspection. They then proceed to the first set of electrically driven cross-cut shears. These are equipped with magnetic handling devices and can cut a cold plate of 43 tons/in^2 tensile strength up to a maximum thickness of $\frac{3}{4}$ in. The shear works on the rocking blade principle, the drive being through reduction gearing to eccentric cams which operate connecting push rods on to the top apron of the shear. The plate stampers are operated by a separate oil–hydraulic system. Scrap is dealt with by a roller conveyor on the outgoing side of the shear which discharges down a chute to a simple aligator scrap shear. The suitably cut scrap is then discharged into one of two pans, the deflecting arm being operated by compressed air.

The plate then moves to the rotary shear. This is a slitter–trimmer combination, the slitter being capable of retraction from the line when not required. The trimmer shear is equipped with 31½-in. dia. blades and can cut plates from 31½ in. wide up to a maximum of 138 in. wide. The entry side is equipped with magnetic handling and clamping devices, and with an indicator for lining-up the plate. Side scrap is dealt with by rotary choppers which discharge on to a conveyor and thence into scrap pans.

Having had its leading and side edges cut, the plate then passes to a second set of cross-cut shears which is a duplicate of the first but has additional equipment in the form of a squaring mechanism and a length gauge. The back-end scrap is dealt with as at the preceding cross-cuts. The plate, cut for length or multiple lengths, then passes on to conventional piling tables.

12-ft roughing mill

The mill is designed to roll 25000-lb slabs having a maximum dimension of 104 in. × 52 in. × 17 in. thick. There is a high-pressure descaling station on the approach table, operation of which is initiated by photocells.

The table rollers are mounted in roller bearings, the slab receiving and approach tables being driven by independent ac motors and the main tables by independent dc motors.

The rolls, which are driven by two 3000-hp motors at 40–80 rev/min, have a 180-in. barrel and a maximum diameter of 46 in. The 31½-in. dia. roll necks are supported by fabric bearings. The whole roll assembly is mounted on a sledge coupled to a duplex roll-change rig. This enables a spare roll set to be built up ready to run into the mill, which is especially valuable during a mid-week roll change when time saved is a direct increase in mill availability.

The mill is equipped with sideguard manipulators on the in- and outgoing sides. Load cells are provided on each screw and the 1500-ton load meters are duplicates of those provided on the 45-in. slab mill.

Boiler-plate normalizing facilities

These were added in 1956–57 and comprise two batch-type heat-treatment furnaces, a cold leveller for 2-in. thick plates, and a cross-cut shear for 1½-in. thick plates.

Frodingham section mills

The conversion and reconstruction of the Frodingham mills was completed in 1949. In 1953 a new stripper shed was built to replace the original; and in 1955 a new roll-turning shop and office was built. In 1956 a new soaker was added. In 1954 production amounted to some 350000 tons/year and is now about 400000 tons/year. The mill layout is shown in Fig.3.

Perhaps the most significant addition in recent years at Appleby-Frodingham is the installation of the universal beam mill now proceeding. While not increasing capacity, it will add the universal range of sections to the long-established B.S. range.

The beam mill consists of a universal roughing group and a single-pass universal finishing stand. The roughing group, a universal and an edging stand, will be supplied with formed blooms from the 32-in. section mill via a skid transfer bank (*see* Fig.3).

The universal stands will have horizontal rolls of a nominal diameter of 52 in. mounted in roller bearings, the vertical rolls being nominally 38 in. dia. The universal roughing mill will be driven by a 5500-hp 0–60–150 rev/min motor through a pinion drive. The edging rolls will have a nominal diameter of 36 in. and will be driven through pinions by a 1500-hp 0–110–275 rev/min motor. The universal finishing stand will have a 2500-hp motor.

After the requisite number of passes in the roughing group, the bars will pass down conveying roller tables into the single-pass finishing stand which will be a duplicate of the universal roughing mill. The three mill stands, i.e. universal rougher, edger, and universal finisher, are all equipped with pre-set screw-down control, and particular attention has been paid to the speed-matching controls on the driving motors.

Sections leaving the single-pass universal finishing stand will be conveyed by independently driven ac-motored rollers to a transfer bank. This will bring the products of the mill into the runout table of the existing section-mill finishing stand. The bars then pass to the hot saws, cooling banks, roller straighteners (for lighter sections), and thence to loading banks. The heavier sections will be straightened and processed on a separate cold bank and gag press installation.

One of the interesting engineering features of the new beam mill will be the methods employed for the disposal of scale. The system will consist of two collecting basins, one at the roughing group and one at the finishing stand. Scale which will gravitate to the bottom of the conical collecting basin will be pumped away to a series of hydro-cyclones graded by size and design to take out the varying sizes of scale. Scale which is too fine to be taken out by the hydro-cyclones will be taken out by conventional sand filters.

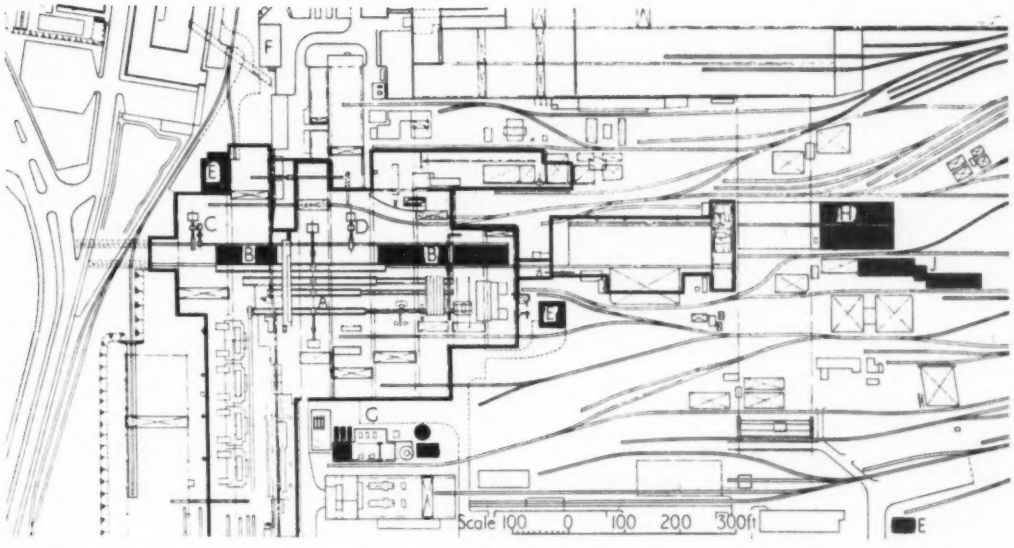
Other developments since 1954

Locomotives and cranes

The original steam locomotive and crane fleet is being replaced by diesels. So far 18 out of 47 locomotives have been replaced and ten out of 30 cranes.

Waste-heat boilers

It is planned to equip all steel furnaces with waste-heat boilers, and so far six have been equipped. When



A Section mills **B** Transfer banks **C** Universal beam roughing mill **D** Universal beam finishing mill **E** Electrical equipment **F** Mill office **G** Water-treatment plant **H** Cold banks **J** Gag press

3 *Frodingham section mills*

the scheme is complete, it is envisaged that all satellite fuel-fired steam-raising plant other than the coke-oven boilers will be closed down. These will be retained to serve the coke ovens and to act as a balancing unit for the whole works.

Air pollution

The situation and work done at Appleby-Frodingham has already been described elsewhere* and it is possible only to touch on the subject here. Smoke emission from boilers is now negligible; emission from locomotives is still the largest source but is being reduced by the changeover to diesels. Dust emissions have arisen in the past mainly from the ore preparation plants, but now melting shops are a potential source. The arrangements on the former are being continuously extended and on the latter have now got beyond the experimental stage. One bag-filter plant and two dry electrostatic plants are in service on Ajax furnaces. Gaseous pollution is being reduced as waste-heat boilers replace fuel-fired boilers, and will be still further reduced when the coke-oven gas desulphurization plant comes into service shortly.

This must be regarded merely as a summary of the subject and for more details reference should be made to published works cited.

Office and amenity buildings

Since 1954 an extension to the general offices has been built. Similarly, offices have been provided for the engineering, production services, education (including training workshops and lecture-room facilities), test house, coke ovens, and section mills departments. New welfare buildings have been built for some 2000 people, including a new medical centre. The central laboratory has been doubled in size.

* S. H. Brooks and W. J. Calvert: *ISI Spec. Rep. no. 61*, 1958, 5–15.

FUTURE DEVELOPMENTS

In the spring of this year (1960), a development plan was approved by the Iron and Steel Board embracing iron- and steelmaking and rolling plant. In broad outline, this plan provides for both iron- and steelmaking capacities to be raised to about 2 m. tons/year, the construction of a continuous casting plant for blooms, and for a billet and rod–bar mill. Brief particulars are as follows:

Ironmaking

Further production-scale furnace trials have revealed that if the furnaces can be charged with a graded sinter, then the existing blast-furnace plant has a capacity of approaching 40000 tons/week if provided with additional stove and gas-cleaning plant capacity.

The ironmaking plan therefore includes:

4 additional sinter machines with 168-ft long × 8-ft wide strands complete with 67-ft mean dia. sinter coolers, return fines conditioning system and improved sinter screening on all plants
3 additional ore driers
A fine ore bedding plant
4 additional blast-furnace stoves
Extensions to gas-cleaning plants
1 additional 135000-lb/h blast-furnace gas-fired boiler
All necessary auxiliary plant and services.

When complete the combined sinter plant capacity will approach 150000 tons/week of which rather more than 110000 tons/week will be available to the blast-furnaces, and 40000 tons/week will be undersized returns for re-sintering.

Steelmaking

It is planned to convert eight OH furnaces to operate the Ajax process and one OH furnace to an active mixer. Three furnaces have already been converted

and when the conversion programme is complete, steelmaking capacity will approach 40000 ingot tons/week. Each Ajax furnace will be equipped with a waste-heat boiler and a fume-cleaning plant.

Continuous casting

It is planned to install a continuous casting plant having a capacity of 5000 tons/week of 9-in. square blooms. This will be a four-strand machine with a maximum casting speed of 60 in./min. The maximum cut lengths will be 24 ft 6 in.

Billet and rod–bar mill

The billet mill will consist of a reversing breakdown mill followed by a continuous billet mill for the reduction of cast or rolled blooms to appropriate billets for the rod–bar mill. The rod–bar mill will have four strands, laying and pouring reels, and provision for a straight cooling bed to be added later.

ACKNOWLEDGMENTS

The authors wish to express their thanks to the directors of Appleby-Frodingham Steel Company for permission to publish this paper, and to colleagues for their assistance in its preparation.

POWDER METALLURGY JOINT GROUP | The Iron and Steel Institute / The Institute of Metals

SUBMISSION OF PAPERS

Discussions at meetings of the Powder Metallurgy Joint Group so far have been restricted to papers invited in connection with symposia or informal discussions on specified themes. However, the Committee hopes that future discussions will be held on papers or groups of papers submitted for publication in the Group journal *Powder Metallurgy* without invitation.

It is thought that members of the Group and readers of *Powder Metallurgy* may have received the impression that only 'invited' papers are desired and published, but in fact the Committee is interested to receive for consideration first-class papers within the scope of the journal, from authors from all parts of the world, so that *Powder Metallurgy* will become a truly international journal.

All papers submitted must pass through the usual refereeing procedures, and although no guarantee can be given that all papers accepted will be discussed, as far as possible papers that might give rise to useful discussion will be included in meeting programmes. Also, the Committee welcomes suggestions for themes of topical interest for discussions at future meetings of the Group.

A leaflet 'Papers for publication in *Powder Metallurgy*' is available from the Editor, 'Powder Metallurgy', 17 Belgrave Square London SW1.

STAL *in English*

No.7 of the 1960 series of *Stal in English* has been published and subsequent issues will appear at about three-weekly intervals. Full details of subscriptions to *Stal in English* are available from the Secretary of The Iron and Steel Institute.

NEWS

Announcements and News of Science and Industry

THE IRON AND STEEL INSTITUTE

AUTUMN GENERAL MEETING 1960

As previously announced, the Autumn General Meeting of the Institute will be held at Church House, Great Smith Street, London SW1, on Tuesday and Wednesday, **29 and 30 November 1960,** beginning at 9.30 am.

The proceedings will include a meeting on fluid flow in furnaces and converters (*29 and 30 November*), and a meeting on martensite (*29 November*). Full details of these meetings were given in the July 'News' (p.341).

SYMPOSIUM ON STEELS FOR REACTOR PRESSURE CIRCUITS

The Institute is also arranging a symposium on steels for reactor pressure circuits. This will be held at the Hoar Memorial Hall, Church House, Great Smith Street, London SW1 from Wednesday to Friday, **30 November to 2 December 1960.** It will be included in the activities of the British Nuclear Energy Conference, of which the Institute is a member. Provisional details of the symposium were given on p.91 of the September 'News'.

IRON AND STEEL ENGINEERS GROUP Meeting in Scunthorpe, 24–27 October 1960

The forty-first meeting of the Iron and Steel Engineers Group will be held in Scunthorpe from Monday to Thursday, **24 to 27 October 1960.** The programme will include technical sessions at which papers on developments at Appleby-Frodingham, lighting in the iron and steel industry, failures in a blooming mill, and redeveloping an integrated iron- and steelworks will be discussed. Preprints of papers to be presented have been distributed to those registered to attend the meeting, and they appear on pp.195–225 of this issue of the *Journal*. The meeting is fully booked, and no further applications can be accepted.

In addition to the technical sessions, there will be visits to the works of the Appleby-Frodingham Steel Company, John Lysaght's Scunthorpe Works Ltd, and Richard Thomas and Baldwins Ltd, by invitation of the companies. The Lincolnshire Ironmasters' Association will be hosts at dinners on the evenings of 25 and 26 October.

The programme of the meeting is as follows:

Monday 24 October

Members arrive at hotels in the Scunthorpe area.

Tuesday 25 October

10 am–12.30 pm
Technical session at the Civic Theatre at which the following papers will be presented and discussed:

'Developments at Appleby-Frodingham since 1954' by *E. A. Atkin and E. F. Farrington* (Appleby-Frodingham Steel Company)

'Investigation of failures in a blooming mill', *by F. E. Stacey* (John Lysaght's Scunthorpe Works Ltd)

12.30 pm–5.30 pm
Works visits (with luncheon) to one of the following:

A *Appleby-Frodingham:* ore preparation, sinter plant, ironmaking

B *Appleby-Frodingham:* steelmaking, rolling mills

C *Lysaght's:* coke ovens, ore preparation, ironmaking

D *Lysaght's:* steelmaking, rolling mills

E *Richard Thomas and Baldwins:* complete Redbourn works

Wednesday 26 October

10 am–12.30 pm
Technical session at the Civic Theatre, at which the following papers will be presented and discussed:

'Engineering aspects of the Redbourn development scheme', *by P. M. Hesketh* (Richard Thomas and Baldwins Ltd)

'Some aspects of lighting in the iron and steel industry', *by D. R. Nisbet* (Appleby-Frodingham Steel Company)

12.30 pm–5.30 pm
Works visits (with luncheon); the same five alternative visits will be offered as on the afternoon of 25 October.

Thursday 27 October

10 am–12.30 pm
Works visits as on the afternoon of 25 October.

1.00 pm
Lunch at the Berkeley Hotel, Scunthorpe.

END OF MEETING

SIDNEY GILCHRIST THOMAS MEMORIAL FUND

The following subscriptions have been received:

	£	s	d
Previously acknowledged	3144	19	0
Stewarts and Lloyds Ltd	500	0	0
Colvilles Ltd	500	0	0
The United Steel Companies Ltd	500	0	0
Lancashire Steel Corporation Ltd	500	0	0
Daniel Doncaster and Sons Ltd	150	0	0
Pickford Holland and Co. Ltd	100	0	0
John Summers and Sons Ltd	100	0	0
Martin Burn Ltd	100	0	0
Dorman Long and Co. Ltd	100	0	0
Ashmore, Benson Pease and Co. Ltd	50	0	0
Union Carbide Ltd	50	0	0
English Steel Corporation Ltd	50	0	0
R. B. Tennant Ltd	50	0	0
London and Scandinavian Metallurgical Co. Ltd	25	0	0
John G. Stein and Co. Ltd	25	0	0
The Patent Shaft Steel Works Ltd	25	0	0
The Llanelly Steel Co. (1907) Ltd	21	0	0
Crofts (Engineering) Ltd	20	0	0
The Incandescent Heat Co. Ltd	10	10	0
Thomas Wragg and Sons (Sheffield) Ltd	10	10	0
Stein and Atkinson Ltd	5	5	0
AEI–Birlec Ltd	5	0	0
Salem Engineering Co. Ltd	5	0	0
H. H. Aston Esq., Newport, Mon.	2	2	0
Total	6049	16	0

In addition to the above, the Verein deutscher Eisenhüttenleute has subscribed 50000 DM on behalf of the German steel industry.

NEWS OF MEMBERS

Mr **W. C. Bell** (*vice-president*) has been appointed to the board of the Wellman Smith Owen Engineering Corporation Ltd.

Mr **R. C. W. Cameron** has been appointed general technical superintendent of Durgapur Steelworks, India.

Mr **D. L. Campbell** has been appointed to the board of the Wellman Smith Owen Engineering Corporation Ltd.

Mr **W. J. Chadder** has been appointed a director of Woodall–Duckham Ltd.

Mr **R. H. Clark** has been appointed honorary secretary of the Refractories Association of Great Britain.

Mr **V. C. Faulkner** has retired from the editorship of *Foundry Trade Journal* to become editorial director.

Mr **C. W. Gunnee** has been appointed manager–technical services of the Redbourn Works of Richard Thomas and Baldwins Ltd.

Mr **R. A. Granville** has left BISRA to join Esso Ltd.

Mr **W. T. Hale** has been appointed assistant managing director of Thomas Marshall and Co. (Loxley) Ltd.

Mr **R. Hardy** is now an investigator in metallurgical studies at Aluminium Laboratories Ltd, Banbury.

Mr **T. Jolly** (*honorary vice-president*), joint managing director of Guest Keen Iron and Steel Company Ltd, will retire from his office and from the board on 1 January 1961.

Mr **J. A. Kilby** has been appointed a director of Colville Constructional Co. Ltd.

Colonel **W. H. Mather** (*honorary member of Council*) has been appointed managing director of the Head Wrightson Machine Company Ltd.

Mr **Joseph Mitchell** has been appointed a director of Walter Somers Ltd.

Mr **W. A. Nicol** has been appointed administrative director of Guest Keen and Nettlefolds Ltd.

Mr **C. F. Pagnamenta,** O.B.E., has been elected a director of Guest Keen Iron and Steel Co. Ltd.

Mr **A. R. Parkes** has been appointed editor of *Foundry Trade Journal*.

Mr **Aksel K. Prestrud** is now acting managing director of Electric Furnace Products Co. Ltd, Sauda, Norway.

Professor **E. G. Rudberg** has been appointed secretary of the Royal Swedish Academy of Science. Professor Dr Roland Kiessling has succeeded him as director of the Metallografiska Institutet, Stockholm.

Mr **L. G. Stock** has been appointed a director (sales) of Richard Thomas and Baldwins Ltd.

Professor **M. W. Thring** is a member of the Advisory Council on Research and Technology to the Minister of Power.

Mr **C. H. T. Williams** (*vice-president*) has been appointed to the board of Tube Investments Ltd.

Mr **R. C. Weir** has been appointed plant and production manager of the new Regina steelworks of the Interprovincial Steel Corporation Ltd, Canada.

Obituary

Harald Biegen (*elected 1947*) of Sauda, Norway, on 9 July.

Professor **Pierre Chevenard** (*elected 1919*), of Paris, on 15 August.

Professor **N. P. Gandhi** (*elected 1935*) of Devlali, India, on 26 June.

Barrington Hooper, C.B.E., (*elected 1932*) of Harrow, Middlesex, on 2 September.

Colonel **Howard Treseder Rees,** D.L., J.P. (*elected* 1952) of Cardiff, on 12 August.

A. F. Tosh, M.B.E. (*elected 1936*), of Low Conniscliffe, nr. Darlington, in 1960.

J. G. Widdowson (*elected 1917*), of Sheffield, on 20 August.

AFFILIATED LOCAL SOCIETIES

Swansea and District Metallurgical Society

Officers 1960–61

President: M. J. A. Thomas, B.SC.
Vice-Presidents: P. Mackay and E. McVicar, F.I.M.
Hon. Secretaries: P. E. Morgan, L.I.M., and R. L. Williams
Hon. Treasurer: L. A. S. Gennett, F.I.M.
Members of Council: Dr C. S. Ball, B.MET., H. A. Cooke, C. H. Cunliffe, M.B.E., D. J. D. Unwin, B.SC., J. D. Richards, B.SC., B. V. Guellard, J. H. Lamb, B.SC., S. M. Ferris, R. H. Humphreys, Prof. A. R. L. Singer.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

BISRA report summary 155

Summary 155, the latest to be issued, deals with the influence of carbon and nitrogen on the magnetic properties of silicon iron, and is available from the Information Section, 11 Park Lane, London W1.

AMERICAN IRON AND STEEL INSTITUTE

Appointments

Mr George S. Rose, Secretary of the American Iron and Steel Institute, is now also a Vice-President of the Institute. Mr Charles M. Parker has been appointed Vice-President–Research and Technology.

INSTITUTION OF MINING AND METALLURGY

President-Elect

Mr A. R. O. Williams, O.B.E., A.R.S.M., B.SC., a managing director of Consolidated Gold Fields of South Africa Ltd, and of New Consolidated Gold Fields, Ltd, has been nominated President of the Institution of Mining and Metallurgy for 1961–2.

THE INDIAN INSTITUTE OF METALS

14th annual technical meeting

The fourteenth annual technical meeting of the Indian Institute of Metals will be held at Bhilai from 1 to 4 December 1960. The programme will include a symposium on the 'Use of oxygen in the manufacture of iron and steel'. The 7th Metallographic Contest and Exhibition will also be held during the session. In addition to the technical meeting, plant visits and social functions will be included in the programme. Bhilai is the location of one of the three new 1-m. ton integrated steelplants which have recently come into production in India.

Members of The Iron and Steel Institute who may be visiting India at the time are invited to participate in the meeting. Further particulars and detailed programme will be sent on application to R. D. Lalkaka, Honorary Secretary, The Indian Institute of Metals, 31 Chowringhee Road, Calcutta 16, India.

THE INSTITUTE OF PHYSICS AND THE PHYSICAL SOCIETY

Symposium on electrical contacts

The Institute of Physics and The Physical Society announces that it is arranging a symposium on 'Electrical contacts' in collaboration with The Institution of Electrical Engineers, which will take place in the Brunel College of Technology, London, on 5–7 April 1961. The symposium is intended to cover recent advances in the study of the phenomena occurring at mating surfaces carrying currents used in light electrical engineering. It will include electrical erosion and material transfer, mechanical wear, and the influence of surface films and contamination, and will deal with make-and-break contacts, sliding contacts, semi-permanent contacts, and connexions between metallic and non-metallic materials. The symposium is intended mainly for those concerned with the design and improvement of contacts. A small informal exhibition will be open during the symposium to those attending the meetings. Abstracts (but not preprints) will be circulated before the symposium, the proceedings of which will not be published in full.

All communications regarding the symposium should be sent to the Secretary, The Institute of Physics and The Physical Society, 47 Belgrave Square, London SW1.

BRITISH WELDING RESEARCH ASSOCIATION

Annual report 1959–60

The report of council of the BWRA for 1959–60 surveys 12 months of varied activity. Welding in recent years has involved the use of new materials and methods which have required extensive research, in addition to the development and improvement of existing techniques, and this report gives a clear idea of the scope of the Association's work. Research is considered under the headings of fatigue investigations, brittle fracture of steel, pressure vessels and pipelines, resistance welding, fundamentals of arc welding, ferrous and non-ferrous metallurgy, and new methods of analysis. Research sponsored by the Association at the universities of Cambridge and Birmingham is also reported.

The report outlines how the results of this research are disseminated, and the extent of the Association's co-operation with other organizations. Finely printed in two colours, the report is available from the Association at 19 Fitzroy Square, London W1.

New appointment

Mr E. H. S. van Someren has recently joined the BWRA as a principal scientific officer. He will work on fundamental studies on the physics of the welding arc by optical observations.

CONTRIBUTORS TO THE JOURNAL

M. A. Glinkov – Head of Department of Metallurgical Furnaces, Moscow Steel Institute (*Contributor to the September Journal*).
Professor Glinkov was born in 1906 and graduated from the Urals Polytechnical Institute in 1927. He gained his degree of doctor of technical sciences in 1941. Since 1945 he has been professor of metallurgical furnaces at the Moscow Steel Institute. Professor Glinkov has published numerous technical articles and several books on the general theory of furnaces and the thermal operation of OH and reheating furnaces. Next month he will be visiting the UK to participate in the Autumn General Meeting of The Iron and Steel Institute.

F. E. Stacey, A.M.I.MECH.E. – Engineer in charge of works mechanical maintenance, John Lysaght's Scunthorpe Works Ltd.
Francis Ernest Stacey studied mechanical engineering at Denbighshire Technical College and the Liverpool College of Technology, and management at the Liverpool College of Commerce. He worked in the general engineering and aircraft industries before joining John Summers and Sons Ltd in 1946 as a draughtsman. Later he became junior maintenance engineer engaged on rolling mills, then assistant engineer at the central engineering workshops. He was awarded a Graduateship of the Institute of Mechanical Engineers in 1951 and an Associate Membership in 1956. Mr Stacey joined John Lysaght's Scunthorpe Works in 1958 as mechanical engineer, and was later appointed to his present post.

F. E. Stacey

P. M. Hesketh

P. M. Hesketh, M.ENG., A.M.I.C.E., A.M.I. STRUCT.E. – Chief engineer, Redbourn Section, Richard Thomas and Baldwins Ltd.
P. M. Hesketh was educated at Liverpool Collegiate School and Liverpool University where he obtained his B.ENG.(HONS.) degree in civil engineering. He joined the Service and Advisory Section of Richard Thomas and Baldwins Ltd at Bridgend in 1948, and in 1950 he transferred to the Redbourn Section as assistant chief engineer. Mr Hesketh took up his present post in 1958.

D. R. M. Nisbet, A.R.C.S.T., A.M.I.E.E. – Chief electrical engineer, Appleby-Frodingham Steel Company, Scunthorpe.
D. R. M. Nisbet was educated at New Cumnock Higher Grade School, and later at the Royal College of Science and Technology, Glasgow, where he was awarded the A.R.C.S.T. in 1938. He then spent some time in the mining industry in Scotland before entering the RAF in 1942. Mr Nisbet joined Appleby-Frodingham as plant electrical engineer in 1947 and was appointed chief electrical engineer in 1959.

D. R. M. Nisbet

E. A. Atkin

E. A. Atkin – Assistant chief engineer, Appleby-Frodingham Steel Company.
Eric A. Atkin was born in 1920 and joined Appleby-Frodingham as an engineering apprentice. After gaining drawing office experience, he held various appointments on the maintenance engineering staff until 1947, when he was engaged on the company's expansion programme. He returned to maintenance work in 1950. In 1955 Mr Atkin served for a year with the general works manager and he was then appointed deputy ironworks engineer. In 1959 he was appointed steelworks engineer and later assistant chief engineer.

E. A. Jenkinson, M.SC., A.I.M. – Senior experimental officer, Metallurgy Division, National Physical Laboratory.
Eugen Alexander Jenkinson attended Hampton Grammar School and later Sir John Cass College, London, where he gained his degrees of B.SC.(ENG.) and M.SC.(ENG.) in metallurgy, from the University of London. He joined the National Physical Laboratory in 1926, and is at present engaged on metallurgical research on the behaviour of metals and alloys under deformation at high temperatures.

L. M. T. Hopkin, B.SC., PH.D., A.R.S.M., A.I.M. – Principal scientific officer, Metallurgy Division, National Physical Laboratory.
Dr Hopkin received his metallurgical education at the Royal School of Mines, Imperial College, London, from which he graduated in 1945. He then joined the British Non-Ferrous Metals Research Association where he studied the

E. A. Jenkinson

L. M. T. Hopkin

creep and fatigue properties of lead alloys, for which he was awarded his PH.D. degree by London University. Since 1952 Dr Hopkin has been with the NPL where he has investigated elastic constants and intergranular fracture of copper alloys. He is at present concerned with the creep properties of materials at high temperatures.

EDUCATION

Corrosion Group Education Panel

The Education Panel of the Corrosion Group of the Society of Chemical Industry has prepared a revised summary of courses including corrosion topics available in universities and technical colleges in the UK in the current session. The summary covers both undergraduate and postgraduate courses, and details facilities for research and the main subjects being studied. The summary is available free from the honorary secretary of the Panel, W. D. Parker, B.SC., Society of Chemical Industry, 14 Belgrave Square, London SW1.

Design for welding

An evening course entitled 'Design for welding in general engineering' has been arranged by the School of Welding Technology of the Institute of Welding at 54 Princes Gate, London SW7. The course, intended for designers and draughtsmen, consists of nine lectures, the first of which, 'Arc welding processes and their relation to design' was given on 18 October. The lectures are held at the Institute at 6.0 pm on Tuesday evenings, and further details are available from the Secretary of the Institute of Welding.

United Steel university scholarships

The United Steel Companies Ltd have announced that they are to offer up to ten university scholarships each year, tenable at any university in the UK. Each scholarship will be worth £500 pa and its value will not be affected by parental income. The scholarships will be offered to boys proceeding to a university to take full-time courses leading to approved degrees in appropriate branches of science or engineering. The company is also prepared to consider a limited number of applications from students of non-science subjects, recognizing that there is a place in industrial management for the good arts graduate.

It is intended that the scholarships should provide industrial as well as academic training, and arrangements will be made for successful candidates to receive industrial experience with the company before going up to the university and also to undertake vacation work in industry. In addition to the scholarship award, students will be paid a subsistence allowance during these periods of industrial training. Applications for the scholarships will be considered from students in their final school year who have achieved the necessary academic standard and have also been offered a place at a university. In addition to academic qualifications, the selectors will be looking for character, initiative, and independence of mind.

On completion of their degree courses, scholars will have the opportunity of joining the company's graduate training scheme, which already has the approval of the appropriate professional institutions. During this period, they will be paid salaries at the current rates for graduate apprentices.

Booklets and application forms are now available on request to the company and enquiries should be addressed to University Scholarships, The United Steel Companies Ltd, The Mount, Broomhill, Sheffield 10.

Union Carbide studentships and fellowships

To encourage and advance research in the field of chemistry, the Chemicals Division of Union Carbide Ltd, which has a major petrochemicals plant located at Hythe, Hampshire, is making a grant of £2150 pa for three years to Southampton University and will later consider an extension of this grant. The grant will provide for a research studentship and a research fellowship in synthetic organic chemicals in the university's Department of Chemistry. The studentship will assist promising students chosen by the university to work for a master's or a doctor's degree and the fellowship should stimulate a research atmosphere.

The Alloys Division of the company is giving similar support to other universities, notably, fellowships of £450 pa to Sheffield University and £450 pa to the Royal College of Science and Technology, Glasgow.

Advances in non-destructive testing

Following the success of the course on 'The techniques of non-destructive testing' held last year, a further course of 10 lectures entitled 'Advances in non-destructive testing' is being held in the Physics Department of the Brunel College of Technology, Woodlands Avenue, Acton, London W3. The lectures, given by well known specialists in the subject, are held on Wednesday evenings from 7 to 8.30 pm, and they began on 28 September 1960. They deal with latest developments in the subject and should appeal not only to graduates and holders of the HNC in engineering and science but also to inspectors and others who have an interest in non-destructive testing.

The fee for the course is £1 and further details may be obtained from the Head of the Physics Department.

FORTHCOMING CONFERENCES AND EXHIBITIONS

Conference on irreversibility and statistical mechanics

The Institute of Physics and The Physical Society announces that it is arranging a conference on statistical mechanics (with special reference to irreversibility). It will be held at Queen Mary College, London, on Monday and Tuesday, 19 and 20 December 1960. A full programme has already been arranged. Topics include the general theory of irreversibility, random processes, liquids (theory and experiment), and irreversibility in gases and in plasma.

Further information and preliminary programmes can be obtained from the secretary, The Institute of Physics and The Physical Society, 47 Belgrave Square, London SW1.

Spanish industrial safety and hygiene campaign

The Centre for Metallurgical Studies and Information in Barcelona is organizing a campaign next month to publicize industrial safety and hygiene both in Spain and abroad. There will be nation-wide exhibitions of safety posters and general safety publications, and publicity will be given to accident and safety statistics. It is also intended to hold conferences and discussions on topics associated with the campaign.

Further details are available from the Secretary-General, CEAM, Avenida Generalisimo Franco 394, Barcelona 9, Spain.

Sir Alexander Fleck to open exhibition

Sir Alexander Fleck, K.B.E., D.SC., LL.D., F.R.S., President of the Society of Chemical Industry, will open the Corrosion and Metal Finishing Exhibition to be held in the Empire Hall, Olympia, London, from 29 November to 2 December.

The opening ceremony will take place at 11 am on Tuesday, 29 November. The exhibition is believed to be the world's largest display of metal finishing products and services. One of the new features of this year's exhibition will be technical meetings arranged by the Plastics Institute and by the Corrosion Group of the Society of Chemical Industry. There will also be technical films dealing with various aspects of corrosion and metal finishing.

NEWS OF SCIENCE AND INDUSTRY

Colvilles appointments

The board of Colvilles Ltd has appointed Mr T. R. Craig, O.B.E., as managing director (sales) and Mr R. P. Towndrow as managing director (works). Sir Andrew McCance, D.L., D.SC., LL.D., F.R.S., will continue as general managing director.

G. A. Smith retires

Mr G. A. Smith, managing director of John Smith (Keighley) Ltd (a subsidiary of Thos. W. Ward Ltd, Sheffield), has, on medical advice, resigned his appointment. Mr Smith is retaining his seat on the board as a director in a consultative capacity.

His successor is Mr Frederick Johnson, a director of the associated company of Thomas Smith and Sons (Rodley) Ltd, who has taken over from Mr Smith as a director and general manager.

New plant and equipment

B. O. Morris Ltd of Coventry, sole UK agents for the Dutch 'Grasso' pneumatic tools, are distributing a new air vice capable of gripping with a force of over ½ ton. Vice jaws are 4½ in. wide and the maximum gap is 5 in. A feature of the vice is that the jaws cannot close on the operator's fingers, since the actual pneumatic movement is only $\frac{1}{16}$ in.

A new company to furnish strip and tin mill equipment of the American Wean design has been formed in the UK under the title of *Wean–Miles Ltd*. The company will assume the manufacture and sale in the UK of many items of equipment designed and sold by The Wean Engineering Company in the USA and throughout the world. Wean–Miles Ltd will maintain the American company's designs while adapting the drawings to British standards. In the case of spare parts intended to replace American parts, the Wean designs will be strictly adhered to, so as to ensure absolute interchangeability.

The head office of Wean–Miles Ltd is at 76 Cannon Street, London EC4.

Italian furnaces for UK

Agreement has been reached between Wellman Smith Owen (Furnaces) Ltd and Leone Tagliaferri AG, Milan whereby Wellmans are now able to offer exclusively in the UK, electric arc melting furnaces and ferro-alloy furnaces of the Tagliaferri design.

Contract news

The Head Wrightson Machine Company Ltd have been awarded a contract worth about £800 000 for a complete seamless tube mill by Tubes Ltd, Desford, near Leicester, a member of the Tube Investments Group. Head Wrightson have also contracted to supply three cooling towers for the new Spencer works of Richard Thomas and Baldwins Ltd at a cost of £180 000.

Ashmore, Benson, Pease and Co. Ltd, a member of the Davy–Ashmore group, are proceeding with the design, manufacture, and erection of a third blast-furnace plant at the Ravenscraig works of Colvilles Ltd. The contract is valued at more than £1 m.

Woodall–Duckham Construction Company Ltd have received orders to build two new coking plants together valued at over £4 m. One is for Colvilles Ltd Ravenscraig works, and the other is for the United Coke and Chemicals Co. Ltd at Orgreave.

New booklet on ISB

'The Iron and Steel Board: what it is and what it does' is the title of a booklet which describes in some detail the composition and work of the Board. Previously this information has not been published in such a comprehensive form, although facets of it have been mentioned in the Board s annual reports and special reports on development. The booklet has been prepared to assist steel users, steel producers, and others generally interested in the steel industry to understand the work of the Board.

Copies are available from the offices of the Board at Norfolk House, St James's Square London SW1.

CHANGES OF ADDRESS

The London office of *Union Carbide Ltd* is now at 8 Grafton Street, London W1 (MAY 8100).

The Scottish office and stockroom of *Martonair Ltd* is now at 128 Nithsdale Road, Glasgow S1 (Pollok 3929).

Flexible Drives (Gilmans) Ltd have moved to Skatoskalo Works, Millers Road, Warwick (Warwick 448–9).

INDUSTRIAL PUBLICATIONS RECEIVED

W. H. A. Robertson and Co. Ltd of Bedford have issued catalogues of their rolling mill equipment and wire-drawing machines, printed in the main European languages.

Heat treatment furnaces is a new leaflet from the Electric Resistance Furnace Co. Ltd.

Wiggin Nickel Alloys no.57 includes articles on the use of Incoloy DS in gas-carburizing furnaces, the metallurgy of spark plug alloys, and solenoid valves.

Leaflet RD95 from the Morgan Crucible Company Ltd deals with 'Purox' oxide refractories.

Demag News no.158 has articles on the rapid changeover of continuous light- and medium-section mills, and 12·5 ton grab-equipped floating cranes with automatic weighing equipment.

The August 1960 issue of the Bulletin from the European Community is a special number entitled 'The common market—a year of consolidation'.

Melting furnaces by Monometer is a new leaflet from the Monometer Manufacturing Co. Ltd describing the company's rotary and rotary tilting furnaces.

DIARY

17–21 Oct. SOCIÉTÉ FRANÇAISE DE MÉTALLURGIE—Journées Métallurgiques d'Automne—Maison de la Chimie 28 bis, rue St. Dominique, Paris 7e.

17–21 Oct. NATIONAL METAL EXHIBITION AND CONGRESS—Philadelphia, USA.

18–19 Oct. VEREIN DEUTSCHER INGENIEURE—International conference on cogwheels and gearwheel drives—Aachen, Germany.

18–20 Oct. NATIONAL SAFETY CONGRESS—Central Hall, Westminster, London SW1.

18–20 Oct. ENGINEERING INDUSTRIES ASSOCIATION—13th Engineering Display—Royal Horticultural Society, New Hall, London SW1.

18–21 Oct. NATIONAL INDUSTRIAL SAFETY, HEALTH AND WELFARE EXHIBITION—Granby Hall, Leicester.

18–21 Oct. FIRST EFFLUENT AND WATER TREATMENT EXHIBITION AND CONVENTION—Seymour Hall, Seymour Street, London W1.

19–26 Oct. INTERNATIONAL CONGRESS AND EXHIBITION FOR INSTRUMENTATION AND AUTOMATION—Düsseldorf, Germany.

20–21 Oct. BRITISH IRON AND STEEL RESEARCH ASSOCIATION—28th Blast-furnace Conference—Scarborough.

24–27 Oct. IRON AND STEEL ENGINEERS GROUP—41st Meeting—Scunthorpe, Lincs. (*see* 'News', p.227).

25 Oct. SOCIETY OF INSTRUMENT TECHNOLOGY—Lecture, *Instrumentation past, present, and future* by L. S. Yoxall—Manson House, Portland Place, London W1.

25 Oct. BRITISH NUCLEAR ENERGY CONFERENCE—Lecture, *Some observations on the design and construction of the* N.S. Savannah, by J. A. Dodd and S. MacDonald—Institute of Marine Engineers, 76 Mark Lane, London EC3.

26 Oct. MANCHESTER METALLURGICAL SOCIETY—Meeting and lecture by Prof. C. R. Tottle (The University, Manchester)—Manchester Literary and Philosophical Society, George Street, Manchester.

26–30 Oct. INSTITUTE OF MATERIALS HANDLING—Conference—Peebles.

1 Nov. EBBW VALE METALLURGICAL SOCIETY–ISI ENGINEERS GROUP—Joint meeting and lecture, *Automation in the steel industry* by P. J. Sutherby, M.A.—RTB Old Lecture Room, Ebbw Vale, 7.15 pm.

1–3 Nov. INSTITUTE OF WELDING—Autumn meeting—54 Princes Gate, London SW7.

2 Nov. INSTITUTE OF PETROLEUM—Lecture, *Applied petroleum research on a vehicle dynometer*, by J. Tubman and J. F. Menger—61 New Cavendish Street, London W1, 5.30 pm.

4–6 Nov. INSTITUTION OF METALLURGISTS—Conference—Grand Hotel, Eastbourne.

7–9 Nov. BRITISH IRON AND STEEL FEDERATION—12th Annual Training Conference—The Old Swan Hotel, Harrogate.

8 Nov. INSTITUTION OF ENGINEERS AND SHIPBUILDERS IN SCOTLAND—Lecture, *Axial vibration in a single-screw ship*, by A. J. Johnson and W. McClimont—39 Elmbank Crescent, Glasgow.

8 Nov. INSTITUTION OF CIVIL ENGINEERS—Lecture, *Some contributions from nuclear power to engineering*, by I. Davidson—1–7 Great George Street, London W1, 5.30 pm.

8–9 Nov. COMBUSTION ENGINEERING ASSOCIATION—Conference—Hotel Majestic, Harrogate.

9–10 Nov. INSTITUTION OF MECHANICAL ENGINEERS—Symposium, *The use of secondary surfaces for heat transfer with clean gases*—1 Birdcage Walk, London SW1.

10 Nov. STAFFORDSHIRE IRON AND STEEL INSTITUTE—Presidential address by F. V. Wright—Training Centre, Stewarts and Lloyds Ltd, Bilston, 7.30 pm.

17 Nov. INSTITUTION OF MINING AND METALLURGY—Meeting—Geological Society, Burlington House, Piccadilly, London W1, 5 pm.

18 Nov. WEST OF SCOTLAND IRON AND STEEL INSTITUTE—Lecture, *Selection, testing and use of steel for Caterpillar tractors*, by W. Sellars and W. J. Cormack—39 Elmbank Crescent, Glasgow.

28 Nov.–2 Dec. NATIONAL EXPOSITION OF POWER AND MECHANICAL ENGINEERING—New York, USA.

29 Nov. SOCIETY OF INSTRUMENT TECHNOLOGY—Symposium, *Nuclear temperature scanning*—Manson House, Portland Place, London W1.

29–30 Nov. THE IRON AND STEEL INSTITUTE—Autumn General Meeting—Church House, Great Smith Street, London SW1 (*see* July 'News', p.341.).

29 Nov.–2 Dec. CORROSION AND METAL FINISHING EXHIBITION—Olympia, London.

30 Nov.–2 Dec. THE IRON AND STEEL INSTITUTE AND THE BRITISH NUCLEAR ENERGY CONFERENCE—Symposium, *Steels for reactor pressure circuits*—Church House, Great Smith Street, London SW1. (*See* September 'News', p.91.)

BRITISH IRON AND STEEL INDUSTRY TRANSLATION SERVICE

The following translations are now available, in addition to those given on page 94 of the September 1960 issue of the *Journal*:

The attenuation of X-rays in the wavelength range $\lambda = 0{\cdot}01$–$10{\cdot}0\ \mu$.
MÜLLER, E. A., *Siemens Messtechnik*, 1956, Feb., 37–38; May, pp.107–110. (£2 15*s*. 0*d*.) [**1589**

The behaviour of sulphur in blast-furnace smelting and ways of reducing the sulphur-content of pig iron.
VOSKOBOINIKOV, V. G., *Stal'*, 1955, July, pp.583–591. (£6 5*s*. 0*d*.) [**1592**

The Lubatti electric process for reduction of ores.
PECORARI, A., *Radex Rund.*, 1954, (4/5), pp.152–164. (£3 5*s*. 0*d*.) [**1598**

Cleaning steelworks fumes.
SEPTIER, L., and LEROY, P., Publ. IRSID, 1958, Series A, Dec., (194), pp.1–47. (£9 10*s*.) [**1647**

High-temperature precision investment castings.
SCHWALBE, R., *Stahl u. Eisen*, Jan. 7 pp.36–45. (£6 10*s*. 0*d*.) [**1692**

Automation in continuous rolling mill trains.
ZWICKY, R. *ETZ A*, 1957, Dec. 1, pp.869–873. (£4.) [**1699**

Spectrochemical analysis of solutions by spark excitation. Application to low-alloy steel, copper alloys, and slag.
LARSSON, A., *Jernkont. Ann.*, 1959, (8), pp.508–522. (£4 5*s*. 0*d*.) [**1703**

Embrittlement of heat-resistant austenitic steels.
KRÜGER, G., *Neue Hütte*, 1959, Dec., pp.725–730. (£6.) [**1727**

Investigations on the embrittlement of a heat-resistant austenitic–ferritic chromium–nickel steel.
SCHUMANN, H., *Neue Hütte*, 1959, Dec., pp.733–738. (£3 5*s*. 0*d*.) [**1728**

Structural changes in steel during resistance annealing.
GRIDNEV, V. N., and TREFILOV, V. I., *Izv. Akad. Nauk. SSSR, OTN Met. i Toplivo*, 1959, (2), pp.62–69. (£5.) [**1732**

The influence of electro-deposited protective coatings on the fatigue strength of steel.
TAUSCHER, H., *Draht*, 1959, Oct., pp.511–516; Nov., pp.571–577. (£7 15*s*. 0*d*.) [**1744**

Study on welded iron and steelworks crane bridges of the box-girder type: a contribution to the calculation of thin-walled box-girders.
SCHINDLER, O., *Stahl u. Eisen*, 1959, Dec. 24, pp.1924–1932. (£6 15*s*. 0*d*.) [**1749**

The operational value of iron ore.
FOURT, P. M., *Ann. Mines* (Paris), 1958, Feb., pp.93–111. (£5 15*s*. 0*d*.) [**1757**

Effect of piercing mill roll variables on the deformation process and occurrence of defects in tube manufacture.
PEJČOCH, O., *Hutnické Listy* (Prague), 1960, (2), pp.157–166. (£8.) [**1774**

Oxide inclusions in iron–chromium–oxygen melts (with discussion).
KOCH, W., *et al.*, *Arch. Eisenhütt.*, 1960, May, pp.279–286. (£4 15*s*. 0*d*.) [**1780**

On the corrosion of silica roof bricks with a low content of fluxing agents.
KIENOW, S., *Stahl u. Eisen.*, 1960, June 23, pp.882–888. (£5 10*s*. 0*d*.) [**1813**

Petrographic factors affecting the reducibility of iron ores: an attempt at parallelization.
SCHENCK, H., and KÖNIGS A., *Arch. Eisenhütt.*, 1960, April, pp.201–213. (£8.) [**1817**

The effect of casting technology and of mould surface coatings on ingot quality.
ŠMRHA, L., *Hutnik* (Prague), 1959, vol.9, (6), pp.190–194. (£5 5*s*. 0*d*) [**1823**

Forged tool steels for hot-work.
WAGENER, P., *Rev. Tech. Luxembourg*, 1958, July–Sept., pp.157–179. (£5 5*s*. 0*d*.) [**1841**

ABSTRACTS OF CURRENT LITERATURE *Iron and Steel Manufacture and Related Subjects* AND BOOK NOTICES

CONTENTS

Note: The letters *a*, *b*, and *c* refer to the three columns on the page concerned.

Mineral Resources	229a	Properties, Treatment and Use of Slags	232c	Welding and Flame-Cutting	239b
Ores—Mining and Treatment	229a	Direct Processes	232c	Properties and Tests	239b
Fuel—Preparation, Properties and Uses	230a	Production of Steel	233a	Metallography	242b
Air Pollution and Smoke	230b	Foundry Practice	234c	Corrosion	244a
Temperature Measurement and Control	230b	Forging, Stamping, Drawing, and Pressing	236a	Analysis	245c
Refractory Materials	230c	Rolling Mill Practice	237a	Industrial Uses and Applications	247a
Iron and Steel, General	231a	Machinery and Services for Iron and Steel Plant	239a	Historical	248a
Blast-Furnace Practice and Production of Pig Iron	231a	Lubrication	239a	Economics and Statistics	248b
				Miscellaneous	248c

* Abstracts, the reference to which is marked with an asterisk, are or are to be made available as translations.

ABSTRACTS

These abstracts are also available on index cards in advance of publication in the *Journal*, each abstract being classified under the Universal Decimal system (UDC). Details and subscription notes can be obtained from the Secretary, The Iron and Steel Institute.

MINERAL RESOURCES

Genesis of marine sedimentary iron ores H. Borchert (*Bull. Inst. Min. Met.*, 1960, **69**, March, 261–279) (90 refs).

Future foreign iron ores F. G. Percival (*27th Blast Furn. Conf.*, BISRA, 1959, Oct., 4–8). **Sources, types and conditions of future home ores** T. B. Adkins and D. R. W. Jones (9–17).

ORES—MINING AND TREATMENT

Influence of petrographic factors on the reducibility of magnetic iron ores. Attempt at parallelisation H. Schenck and A. Königs (*Arch. Eisenhütt.*, 1960, **31**, April, 201–213) Investigations are described which have the object of relating the structure of magnetic iron ores to their reducibility, with particular reference to the varying reduction rates of magnetite and hematite, attributable to their lattice relationships (31 refs).

Ore handling and beneficiation at Great Lakes steel G. R. Hornke and R. E. Powers (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 37–60) Improved facilities for burden preparation are expected to increase iron production by 20–30% at the Zug Island plant of Great Lakes Steel Corp. The new ore-handling, screening, crushing, and sintering facilities are described in detail, and their operation to data is discussed.—G.F.

Selection of the optimum economical limit of enrichment of iron ores from the Gorna Shora mines N. S. Sachko, I. G. Mikhailov, and N. A. Shriro (*Izvest. VUZ Chern. Met.*, 1959, (9), 179–187) The principal supplier of iron ores for the KMK blast-furnaces are the Gorna Shora mines with 6×10^6 tons of ore containing $2{\cdot}3 \times 10^6$ of iron. By increasing the limits of enrichment the losses of iron and the production costs increase more than the productivity of the furnace. On this basis the maximum enrichment limits are determined for these mines.

Modern methods of roasting iron pyrites J. Morath (*Tek. Tidsk.*, 1959, **89**, Oct. 9, 965–972) The design, characteristics, and operation of various types of furnace are described and compared.

Hydrometallurgical extraction of copper and cobalt from the iron ores of the Teshan Metallurgical Combine (China) A. L. Zagvanskii and L. A. Petrov (*Izves. AN, USSR, OTN, Metallurg. i Topl.*, 1959, (6), 8–15) The ores are roasted in a furnace at 600° with a fourfold air excess for 4 h. Leaching is carried out four times in a 2–3% H_2SO_4 solution countercurrent, between 60–70°. With a grain size of up to 6 mm the extraction of Co equals 65–76% and that of Cu 75–90%. With a grain size up to 3 mm the maximum extraction increases by 4·6 and 4·1% respectively, and with a grain size below 1 mm by 6·4 and 12%.

Thermal beneficiation of low-grade chrome ores (*Nat. Met. Lab. Ann. Rep.*, 1958–59, 24) This note describes experiments carried out to upgrade the ore by preferential reduction using gaseous and solid reduction methods.

Thermal beneficiation of low-grade manganese ores (*Nat. Met. Lab., Ann. Rep.*, 1958–59, 24) Studies were undertaken to upgrade Mn ores by preferential reduction of iron, producing a Mn-rich slag. The best result in larger scale experiments was obtained by smelting the ore with 7·5% coke dust for 3½ h which gave a slag with 46·97%Mn, 5·26%Fe and a metal with 20·35%Mn and 71·08%Fe, and a recovery of 95%Mn in the slag with a Mn–Fe ratio of 8:9.

Semi-pilot plant studies on the beneficiation of ferruginous manganese ore from Joda West Mine of Tata Iron and Steel Co. Ltd (*Nat. Met. Lab., Ann. Rep.*, 1958–59, 35) This note describes a process based on low-temp. magnetizing gaseous reduction which produces high-grade Mn concentrates from ferruginous low-grade Mn ores at an economic cost. Three t of low-grade ore produce one of Mn concentrate. Reference is made to proposals for a plant to beneficiate 600 t/day of this ore. **Pilot plant studies on the beneficiation of low-grade ferruginous manganese ore from Siljora–Kalimati Mines, Orissa** (35) Proposals for a 100 t/day plant are noted.

Pilot plants for the agglomeration of iron ores on a semi-industrial scale H. Rausch and F. Cappel (*NML Tech. J.*, 1960, **2**, Feb., 27–34) A variety of tests is discussed and criticized, and investigations of sintering and pelletizing on industrial and semi-industrial scales are compared. The complexity of the subject is stressed and many unsolved problems are indicated.—S.H.-S.

Industrial trials with Vol'kov iron–vanadium concentrate N. F. Dubrov (*Stal'*, 1960, (5), 405) A note from the Ural Scientific Research Institute.Trials on the sintering, blast-furnace smelting, and converter treatment of the concentrate are briefly noted. Steel with 0·03–0·5%V was obtained in 17–20-t charges and V was recovered from the slag with 10–17% V_2O_5.

Production of self-fluxing pellets in the laboratory and pilot plant K. E. Merklin and F. D. DeVaney (*Min. Eng.*, 1960, **12**, March, 266–271) Tests on the pelletizing of taconite are reported, especially with additions of dolomite. Structures and composition of the fired pellets are given. Blast-furnace results are not yet available.

Mixing and conditioning of sinter-plant feed G. Brandes and H. Rausch (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 232–249) The authors stress the importance of correct preparation of sinter mixes, and describe the types of mixers and granulating equipment currently used. The conditions under which re-rolling after mixing is desirable are discussed, and test results are given to show the influence of mixing on several typical sinter mixes. It is concluded that no one optimum mixing process can be applied to every kind of sinter mix.—G.F.

Control and proportioning of sinter plant raw materials J. W. Meredith and A. M. Frankau (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 60–67) The authors emphasize the necessity for accurate proportioning of the ingredients of a sinter mix, and describe the Huntington Heberlein control system which ensures the required accuracy with a high degree of reliability.

The making of self-fluxing sinter and the blast-furnace practice with its 100 per cent sinter burden Y. Doi and K. Kasai (*AIMME, Blast Furn Comm. Proc.*, 1959, **18**, 182–205) A self-fluxing sinter practice has been used for about a year at Kokura Steel Works, and a blast-furnace has been operated successfully and efficiently on a burden of 100% self-fluxing sinter. The preliminary development work is described, and details are given of the operating practice.—G.F.

Mechanism of mineral formation in the sintering of fluxed sinter D. A. Kissin and T. I. Litvinova (*Stal'*, 1960, (5), 397–403) Petrographic, magnetometric, and chemical analytical studies of experimental sinters were made to identify the constituents formed and to elucidate their mechanism of formation. Formation of fayalite, olivine, silicates, and ferrites are considered and trials with and without flux were made. $CaFe_2O_4$ appeared to have the best metallurgical properties. The strength and reducibility of this and other constituents were studied.

Sintering studies on the magnetic concentrate produced from low-grade Salem magnetite ore (*Nat. Met. Lab. Ann. Rep.*, 1958–59, 15) The results of an investigation on the effects of variables such as basicity, coke, and moisture contents in sinter mix on the quality of sinter produced carried out on a laboratory sintering unit showed that the optimum amounts of coke and water contents were 7·0 and 9·0% for self-fluxing sinter with a basicity ratio of 0·85.

New results with iron ore sintering, using blast-furnace gas as an additional fuel J. Holzhey (*Neue Hütte*, 1960, **5**, March, 143–148) The process is described and results compared with those from normal sintering operation. Advantages include greater uniformity of

temp. development in the sinter bed; temp. was held longer in the upper layers resulting in an increase in sinter strength, without loss in sinter reducibility.

Scottish experience with fluxed sinter E. H. Baldwin and I. M. Mathieson (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 218–231; *27th Blast Furnace Conf. BISRA*, 1959, Oct., 24–33) The authors describe the theoretical considerations and the subsequent experiments which led to the adoption of fluxed sinter at Colvilles Ltd. Typical operating results at Clyde and Ravenscraig Works are given.—G.F.

Production and use of ferro-coke G. Granger, M. J. Lack, and J. Mitchell (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 152–161) The production of ferro-coke by incorporating blast-furnace flue dust into a carbonizing mix, and its use as furnace fuel, is a simple and economical means of recovering the iron and carbon in flue dust. The authors describe the development of the operation at the Everett plant of Eastern Gas and Fuel Associates.

Ferrocoke from lignite and magnetite ores of South India H. C. Nandi and M. S. Iyengar (*J. Sci. Ind. Res.*, 1960, **19A,** Feb., 84–87) The production of satisfactory briquettes is possible.

Basic principles of the manufacture of sponge iron S. Eketorp (*Tek. Tidsk.*, 1960, **90,** March 11, 245–246, 247–248, 249–250) The chemistry of various direct reduction processes is outlined briefly.

FUEL—PREPARATION, PROPERTIES AND USES

Canadian coals in ferrous metallurgy A. Ignatieff, J. C. Botham, and J. H. Walsh (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 136–150) The quality of Canadian coal reserves is reviewed, the differences between Eastern and Western Canada being described. The coking properties of several coals are tabulated, and the prospects for the use of Western Canadian coal in the ferrous industry are discussed.—G.F.

Theoretical principles of flotation and selection of reagents for coal flotation and some ores difficult to float I. N. Plaksin (*Rudy i Metale*, 1960, **5,** (3), 86–93) In the USSR Academy of Science, tests have been carried out to establish the principles of reagents election for coal flotation and for some ores which are difficult to float. The non-polar reagents are effective with lustrous components of coal slightly oxidized. The action is of a collecting type. Heteropolar substances are effective for dull ash-coals; they have mainly a frothing function. Aliphatic alcohols are satisfactory reagents. The most active are the alcohols containing 6 to 8 atoms of carbon. Tests with W and Nb ores and some sulphides have shown the advisability of using reagents which act differently on minerals of differing crystallographic structures. In certain cases the use of gas is necessary to separate minerals with very similar flotation characteristics.

Better coal preparation through better crushing A. H. Brisse and B. R. Kuchta (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 17–21) The authors discuss the need to improve the sizing of coal charged to coking batteries, and consider the possibility of satisfactorily attaining high levels of pulverization through the use of crushers of improved design.

New developments at Pennsylvania crusher division J. J. Bradfield and F. A. Fawcett (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 22–29) The authors compare the efficiencies of attrition crushing and impact crushing, and outline the design and operation of the Pennsylvania reversible hammermill and reversible impactor. The concept of impact crushing without retention is discussed.—G.F.

Advances in impact crushing techniques as related to coal J. C. F. Taylor (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 29–30) The author briefly discusses recent advances in the design of impact crushers which result in better product control and a reduction in the amount of fines produced.—G.F.

Preparation of coking coal with the Joy-Hazemag impact crusher M. B. Mairs (*AIMME Blast Furn. Comm. Proc.*, 1959, **18,** 31–33) A description is given of the Joy-Hazemag impact crusher which is designed to use, as far as possible, only the impact principle for reduction and to be able to perform the work without oversize recycling. Results indicate that pre-screening can be eliminated, and that the impactor can be used as a coal mixer.—G.F.

Advances and experiences in the evaluation of coal and coal mixtures in the production of metallurgical coke M. F. Kessler (*Hutn. Listy*, 1960, **15,** (1), 6–12) Cokeability and its determination are discussed with special reference to data obtained in Czechoslovakia. The international classification code is shown to be also applicable to mixtures of coals.—P.F.

Automatic sampling and bulk density control of coking blends S. Prellwitz (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 4–13) The author explains the importance of the bulk density of a coking blend, and describes the system used at Fairless Works of United States Steel Corp. where bulk density is continuously measured, and controlled by varying the oil additions. Operating results are compared with those obtained before the control was used.—G.F.

New insight into the coking processes D. W. van Krevelen (**Brennstoff-Chemie*, 1956, **37,** 101–106) The mechanism of the softening and degasifying of bituminous coal was investigated, using a plastometer, dilatometer, and thermobalance. It is concluded that the softening of bituminous coal is based on a type of depolymerization, with simultaneous cracking of the resulting metastable transition products; these act as softeners of the coal.

Coking mixtures with additions of low quality coals J. Kunc and V. Stuchlik (*Paliva*, 1960, **40,** (3), 74–77) Researches were carried out on six 'economical' mixtures containing various amounts of low-grade coal, to ascertain the cokeability. The work is designed to meet the situation expected to arise in a few years when high-grade coking coals will become scarcer. Conditions under which the low-grade coals can best be utilized are indicated; the high-quality coal must not be present in amounts less than ~20%.—P.F.

Methods of banking coke ovens A. D. Shattuck (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 166–173) The author briefly reviews methods of banking coke ovens and discusses the advantages of pressurization. The Clairton pressurizing system is outlined and then details are given of the method used at Great Lakes Steel Corp.—G.F.

AIR POLLUTION AND SMOKE

Recommendations on heights for new industrial chimneys G. Nonhebel (*J. Inst. Fuel*, preprint, 1960, May 23, pp.17).

Dust technology (Annual review) W. Gründer (*VDIZ*, 1960, **102,** May 21, 629–633) A review of recent literature dealing with dust measurement, morphology, and industrial control (54 refs).

Trends in the technique of dust removal in workshops W. Muhlrad (*Usine Nouv.*, 1959, Monthly issue, Nov., 53, 55, 57, 59–61) A review of the various types of installations with cost comparisons practical points on design and installation and of dust measurement.

Cleaning gases at high temperatures W. Strauss and W. M. Thring (*Engineering*, 1960, **189,** April 22, 558) Filter beds for use above 600°C are considered, including use with an OH furnace.

TEMPERATURE MEASUREMENT AND CONTROL

Two colour pyrometer tells correct temperature Shaw Instrument Co. (*Steel*, 1960, **146,** April 4, 144–150) The operating principle and applications of the 'Shawmeter' pyrometer are described. The instrument measures temp. above 1400°F by measuring the ratio of radiation at two selected wavelengths; the same photosensitive element is used for both wavebands in rapidly alternating sequence. Thus it is relatively independent of the emissivity of the object, and the readings are not seriously affected by distance or intervening haze and smoke. It has been used for many metallurgical applications.—D.L.C.P.

Thermocouple materials for the measurement of temperatures above 1600°C J. T. Miller (*Instr. Practice*, 1960, **14,** May, 510–513) A review of combinations of Pt-group, Mo, Re, Ta, W metals and graphite and carbide combinations.

The behaviour of the Venturi pneumatic pyrometer in industrial furnaces R. E. Holland R. Jackson, and G. G. Thurlow (*J. Inst. Fuel*, 1960, **33,** April, 180–187) The design and principles of the venturi pneumatic pyrometer are described and the results obtained in using the instrument in a number of industrial appliances are discussed. These include the OH furnace, Bessemer converter, glass tanks, oil, and pulverized fuel-fired water tube boilers. Finally tests carried out in the experimental pulverized free-fired furnace at IJmuiden using both the pneumatic and suction pyrometers are described. These show the advantage of the former in that it is equally accurate at all temp., has a rapid response, blocks only slowly in dirty gases, and is easily cleaned.

Measurement of very high temperatures A. S. Kaufman (*Instr. Practice*, 1960, **14,** May, 502–507) Plasma conditions and ion electron temp. measurement up to 10^5°K are discussed.

Do you really know how hot it is? (*Steel*, 1960, **146,** March 14, 94–99) The principles of temp. measurement are reviewed and measuring instruments are described with a view to indicating the field of application of each.

Control of furnace temperatures W. J. Swinn (*Vitr. Bull.*, 1960, **10,** April, 327–340, discussion, 340–343) A general account of feed-back control systems.

Experience with continuous temperature measurements of liquid steel in ladle and furnace P. Ya. Sorokin (*Metallurg.*, 1959, (12), 19–21) The Metallurgical Institute of the Ural Branch of the Academy of Science USSR evolved a method for the production of thermocouple tips of ZrO_2 of high thermal stability and high heat resistance. The heat inertia through a 3 mm thick wall does not exceed 60–70 sec. which is of no importance in continuous measurement. The hot junction of the thermocouple is fixed in the ladle through a false stopper, in the bottom or at the sides. It was found that the temp. of the metal in the upper layers of the ladle is less than that towards the bottom. This system of continuous measurement should prove of great advantage when automating production.

REFRACTORY MATERIALS

Researches on the production technology of high-quality compact fireclay blocks F. Tomšů (*Hutn. Listy*, 1960, **15,** (1), 40–46) The technology of production of firebricks containing about 40% of Al_2O_3, and having a porosity of less than 14% was developed. Details of the method and of tests carried out with it on a semi-works scale are given. The process will be introduced on a large scale in Czechoslovakia in the near future.—P.H.

Problems of recent refractories engineering Y. Kora (*Tetsu to Hagane*, 1959, **45,** Dec., 1363–1373). A review—K.E.J.

Evaluation of clinkered dolomite materials A. Sontz and R. N. Ames (*AIMME, Proc. OH Comm.*, 1959, **42,** 165–175) As a result of laboratory tests, subsequently confirmed in OH furnace trials, the authors have established a correlation between the grain size of burnt dolomite and its three most important physical properties, the angle of rest, bulk density, and sintering tendency. The quality of a chemically uniform burnt dolomite can thus be determined merely from sieve analysis.—G.F.

The effect of TiO_2 on the properties of silica refractories N. F. Eaton and R. Higgins (*Trans Brit. Ceram. Soc.*, 1960, **59,** Jan., 1–15) TiO_2 affects the α–β cristobalite transformation. It also affects porosity and permeability. A mechanism for the effect is suggested.

On the formation of tridymite: The joint action of lime and alkaline oxides W. L. De Keyser and R. Cypres (*Silicates Ind.*, 1960, **25,** March, 109–137) A detailed study has been made of the effect of small quantities of the oxides of Na, K, and Li on the polymorphous transformations of silica under the action of CaO.—P.F.

Reaction between molten steel and refractories. IV: Behaviours of C, Si, and O in the steel melted in Al_2O_3–SiO_2 crucibles in a

vacuum S. Minowa and M. Kato (*Rep. Govt. Ind. Res. Inst. Nagoya*, 1960, **9**, Feb., 79–84) After 5–20 min contact at 1600°C Si and O increased in steel melted in pyrophyllite crucibles and O_2 decreased in high-Al_2O_3 crucibles *in vacuo*. C decreased in both. The CO pressure in molten steel is inferred.

A device for firing electric furnaces automatically G. H. B. Lovell (*Trans. Brit. Ceram. Soc.*, 1960, **59**, Jan., 22–24) As described, the device is used for the firing of ceramics.

IRON AND STEEL, GENERAL

The most important scientific and didactic attainments of the Akademia Gornieso–Hutnicza in Cracow and research trends in iron metallurgy W. Rozanski (*Hutnik*, 1960, **27**, (3), 92–95) The author gives an outline of the more important work done. Most of it is of a theoretical nature and can be placed under the following sections. The metallurgy of iron, steel, and alloy steels, the plastic processing of steel, heat treatment of steel, the economies of heat utilization and the building of furnaces and coke ovens.

Developments at Appleby-Frodingham since 1954 E. A. Atkin and E. F. Farrington (*JISI*, 1960, **196**, Oct., 217–224) [This issue].

Material economy in ferrous metallurgical operations L. Mayre (*Neue Hütte*, 1960, **5**, April, 202–212) A detailed study of material economy in iron- and steelmaking operations is made, with particular reference to West German practice and its application to conditions in East Germany.

Engineering aspects of the Redbourn development scheme P. M. Hesketh (*JISI*, 1960, **196**, Oct., 213–218) [This issue].

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Quo Vadis blast-furnace? M. O. Holowaty and J. B. Austin (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 310–314) The authors enumerated the problems which must be overcome if the blast-furnace process is to keep pace with the rapidly progressing steel industry. They consider the possibility of decreasing coke costs, and discuss problems of burdening, furnace design, refractories, and the condition of the blast.—G.F.

New blast-furnace design boosts iron yields (*Iron Age*, 1960, **185**, Jan. 28, 96–98) Proposals by the Koppers Co. Inc., for an 'ultrahigh-pressure' blast-furnace are considered. Reduction rates increase with pressure, and a Freyn rating of 200% is expected. Features of the design are: top pressure 40 psi; 28 ft hearth dia.: two tapholes: fully prepared burden: reduced stack height: gas cleaning effected by a small dust catcher and wet cleaning by orifices in downtake gas main: specially designed blowers. A table summarizing the economics, shows that compared with a normal-pressure furnace of 2000 t/day capacity, high-pressure could double the output and save $\$4 \times 10^6$/year for 23% extra capital cost.—D.L.C.P.

Ultrahigh pressure stack boosts hot metal output O. R. Rice (*Steel*, 1960, **146**, March 21 148–154) The proposed design for a blast-furnace operating at 40 psi top pressure is outlined. A Freyn rating of 200% should be possible; the increased capital cost would be recovered by decreased operating costs. Features include: special charging mechanisms; reduced stack height; external water cooling of stack and hearth; air-cooled base; two tapholes; small wet type dust-catcher and orifice type gas-cleaning plant; fully prepared burden.—D.L.C.P.

New ways of intensifying metallurgical processes I. P. Bardin (*Izvest. AN USSR, OTN, Metallurg. i. Topl.*, 1959, (6), 3–7) A short program outlining the intended development in the production methods such as direct production of steel in blast-furnaces, increasing the size of metallurgical units, improvement of production indices, enrichment of ores, increasing hearth temperatures by about 200°, reconstruction of the hearths of blast-furnaces etc.

The uses and limitations of pilot plants in studying iron making problems E. W. Voice (*NML Tech. J.*, 1960, **2**, Feb., 24–26) The limitations of pilot plants in processes such as blast-furnaces, where scaling is involved, are compared with those in, e.g. sintering, where it is not. Problems are indicated and discussed.

Ore and coke in blast-furnace models T. Turnbull and J. B. Wagstaff (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 120–134) The authors discuss and illustrate the results of blast-furnace model studies of solid movement, covering the use of ore and coke charges. The movement of columns of different bulk density is described, and several major effects of gas flow are noted.—G.F.

Bureau of Mines use of natural gas in an experimental blast-furnace N. B. Melcher (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 69–73) In experiments on the experimental blast-furnace at Bureau of Mines, natural gas has been injected into the smelting zone through auxiliary tuyeres in quantities up to 8% of the air blast. The coke rate was reduced from 1494 to 948 lb/t, and pig-iron production rate was increased by 30% with no increase in wind rate.—G.F.

Technical achievements of blast-furnace production in China Din Shu-lyan (*Izvest. VUZ Chern. Met.*, 1959, (8), 17–19) Blast-furnace production technique has improved continuously during the last ten years. The best volume coefficient of production was 0·67 in 1958. Soviet Russia is helping very much and modern methods of driving furnaces are being rapidly introduced (full blast, rapid charging, fluxed sinter, etc.). The improvement in the quality of sinter now makes possible the production of low-silicon conversion iron at the Anshan Combine, without needing refining before conversion.

Bending and the rigidity of valve component covers free or flexibly suspended, suitable for blast-furnace equipment D. A. Storozhik (*Izvest. VUZ Chern. Met.*, 1959, (9), 157–169) In blast-furnaces operating with increased top pressure, blow-throughs occur at the valve joints of the equipment. The article presents a detailed study of all components of this equipment and gives suggestions for their construction and joining to all gas fittings.

Standardization of blast-furnace production E. S. Glikman and I. A. Britvin (*Izvest. VUZ Chern. Met.*, 1959, (8), 171–177) The calculation of the efficiency (P) is made from the expression $P = V_e \frac{I}{K}$ where P is the furnace output per 24 h in t, V_e the effective furnace volume in m^3, I the index of melting intensity, i.e. the ratio of the coke consumption in 24 h to the effective furnace volume in t/m^3, and K the specific coke consumption in t per t of iron. The effect of all these terms of the equation is then considered in detail and standard norms are evolved.

Testing lumpy ore for its behaviour in the blast-furnace H. Birnbaum, L. von Bogdandy, and W. Janke (*Stahl Eisen*, 1960, **80**, June 9, 781–788) Studies of the mechanical properties of lumpy iron ore as well as its reduction rate in the blast-furnace showed that certain coefficients could be derived that enable one to predict the behaviour of the ore in the furnace. In this connexion, a formula was also developed for the prediction of the coke consumption in the blast-furnace. Agreement between theory and practice—tested over six months—was good (60 refs).—T.G.

Correlation between preliminary processing of blast-furnace burden materials and low coke rate K. Tsujihata, K. Kodama, and M. Inoue (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 205–215) At Yawata Iron and Steel Works, the preliminary processing of the burden materials, using high-efficiency crushers, screens, and sintering plants, has reduced the coke rate from 1640 to 1200 lb/t. Details of the practice and operating results are given.

Testing a carbon lining for blast-furnace stacks F. A. Khil'kevich, B. L. Lazarev, S. V. Bazilevich, and V. K. Kornev (*Stal'*, 1960, (5), 385–391) Early tests in which the complete bosh was lined and failure was rapidly brought about by Zn compounds and steam and later successful trials on lining the bottom half of stack and bosh with asbestos insulation to remove the zone of zincite crystallization beyond the carbon zone are described. Wear is still considerable and the production of uniform fine-grained blocks of low porosity is called for.

The influence of blast temperature on blast-furnace operation N. N. Babarykin and F. A. Yushin (*Stal'*, 1960, (5), 391–396) Blast temp. has been raised to 1000° and is to be increased further to 1100–1200°. Furnace operations were studied to trace the effects of the increase and temp. and pressure drop were measured, also slag compositions and degree of reduction. Coke rate was reduced, CO_2 was slightly increased, and temp. and pressure changes were noted.

Experimental smelting of a siliceous iron ore H. U. Ross (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 106–119) The author describes experimental work with a siliceous iron ore which could not satisfactorily be concentrated by conventional means. The ore was concentrated to the highest grade possible, sintered, and smelted in an electric furnace under an acid slag. The results are discussed in detail and the practicability of the process is considered.—G.F.

Relation between temperature and composition of pig iron and slag V. I. Loginov (*Izvest. VUZ Chern. Met.*, 1959, (6), 15–21) Investigations are reported concerning the variations of the Si content in conversion iron as related to the actual temp. of iron and slag and concerning the relation between the reduction of Si and the basicity of the slag. The investigations showed that with the changing conditions of blast-furnace production (increased pressure, humidity of blast, fluxed sinter, production of low-Mn, and low-Si irons, etc.), the qualitative and quantitative relations between temp. and composition of the pig iron and the vaporization of the sulphur in blast-furnace production also change.

Selection of conditions for high driving rate of large blast-furnaces of the NTMK N. F. Dubrov (*Stal'*, 1960, (5), 405) A note. Productivity was raised 7% by increasing the sinter charge to 100% and the basicity from 0·75 to 1·15 and coke rate was reduced 8%. Greater blast volume also reduced the coefficient to 0·69–0·73.

Economics of the use of pellets in blast-furnace production I. S. Kagan, Kh. P. Ziatsev, N. S. Snagovskaya, E. N. Volkova, and I. A. Volokh (*Izvest. VUZ, Chern. Met.*, 1959, (6), 145–154) Technical and economic considerations is given to the production of pellets from magnetite concentrates, and to their use in blast-furnaces. The conclusions are that it is economic to replace solid fuel (coke fines) by gas for pelletizing; that pelletizing offers great metallurgical advantages (good reduction possibilities, good gas permeability, uniformity of size etc.); and the great strength of the pellets which allows for long transport without crushing into dust.

Blast-furnace results with pellet and sinter burdens J. B. Hill and H. Epstein (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 177–181) At Bethlehem Steel Co., highly prepared burdens consisting of 90% of all ore charged as pellets, sinter, or screened ore have been used. Results are presented which show the increased iron tonnages and decreased coke rates obtained.—G.F.

Blast-furnace sinter related to burdening and charging F. R. Kik (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18**, 100–105) The author describes his experience with high sinter burdens and discusses some of the problems encountered. He deals primarily with the physical aspects of the sinter burden and with the changes necessary in order to operate the furnace most efficiently.—G.F.

Manufacture of pig irons for nodular graphite cast iron in the blast-furnace and properties of the irons I. Aoki (*26 Congres. Internacional de Fundicion IV. Asamblea General del IHA, Madrid*, 1959, ref.C5, pp.6) [In English] In the production of pig iron at Kamaishi, Japan, Iron Works for the manufacture of nodular cast iron, unsatisfactory casts were sometimes made which were within specification for C, Si, Mn, and S. This was attributed to the presence of Sn, As, Pb, and Sb in association with 0·05–0·1%Ti.

Operating experience of a blast-furnace with evaporative cooling P. G. Glazkov and D. E.

Krasnozhen (*Stal'*, 1960, (5), 403–405) A review of the working of a 1033 m^3 furnace since Nov. 1957 is presented. Plans of the cooling arrangements are given and the results show that they require simplification and improvement.

Controlling the heat balance of the blast-furnace by the analysis of the flue gases A. N. Pokhvisnev and N. K. Zhilkin (*Izvest. VUZ Chern. Met.*, 1959, (9), 29–45) A method is described for maintaining the H_2 balance in the flue gases which permit the quantitative determination of O_2 taking part in the reducing processes and also the volumes of O_2 extracted from the charge by H_2, and leaving the furnace in the form of steam. This is of special importance for the blowing of generator gas. The general formula evolved for a rapid calculation of the heat balance permits at any moment a correct assessment of the heat supplied per unit of oxygen extracted from the charge (or its equivalent per unit of molten iron) and may be used efficiently for controlling the heat balance of the furnace.

The course of reduction and the temperature profile in the blast-furnace E. Schürmann, W. Zischkale, P. Ischebeck, and G. Heynert (*Stahl Eisen*, 1960, **80,** June 23, 854–861) The temp. prevailing over the whole length of a blast-furnace were measured with the aid of probes, at the same time the composition of the gases was determined. The evaluation of the results showed a close relationship between temp., gas composition, and Boudouard reaction. The curves obtained in this investigation show no agreement with the concepts held so far about the reactions taking place in the blast-furnace. The main finding may be summarized as follows: The reduction of the iron oxides in which the CO_2 formed does not react with the coke is confined mainly to a temp. region between 800 and 1000°C.—T.G.

Some studies on the behaviour of gas in blast-furnaces Y. Imao and S. Mizuno (*Tetsu to Hagane*, 1959, **45,** Dec., 1321–1327) A German-type gas sampler was used in the shaft of No.1 furnace of Kokura Works, and a simplified type in No.2. The distribution of CO_2, gas temp., and velocity were different for the two furnaces. Changes in burden influenced the gas behaviour and reducibility: the latter was higher with self-fluxing sinter than with raw ore. About 25% of CO at the sampling level was used in carbon deposition, and 5% in reduction. The CO_2 in ascending gas increased during hanging.—K.E.J.

The technology of producing foundry iron in the blast-furnaces of the Kuznetsk Metallurgical Combine N. N. Chernov, P. G. Zhigulev, A. F. Egorov, and M. D. Karachentsev (*Izvest. VUZ Chern. Met.*, 1959, (8), 21–29) The heating up and cooling of the furnaces must be carried out evenly. Rapid transitions from the melting of conversion iron to that of foundry iron and vice versa are causes of the formation of incrustations in the furnaces. A campaign for foundry iron must not last less than 10–12 days. A shorter period does not give sufficient time to organize a smooth transition. The increase or decrease of the ore burden during the transition must be done by changing the weight of the coke charge without a change in the optimum weight of the ore additions in the charge. Under conditions obtaining at Kusnetsk, the slag must amount to about 0·55 t per t of iron and have a basicity of 0·95–0·97. The quantity of slag increases with the production of hot types of iron.

Results of campaigns carried through in 1958 in the low-shaft blast-furnace of Liège E. P. Bonnaure (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 75–91) The author describes the Liège low-shaft blast-furnace and presents results which illustrate its use as a small experimental furnace. The influence of high top pressure, driving rate, and burden granulometry has been investigated, and successful trials performed with fuel injection and O_2 enrichment of the blast. The work has led to a better understanding of the physico-chemical and thermal processes that take place in the stack.—G.F.

Laboratory evaluation of performance characteristics of various types of sinter A. M. Schwartz and M. O. Holowaty (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 92–98) In order to simulate conditions within the blast-furnace, the authors have developed a laboratory high-temp. blast column and a standard testing procedure for evaluating sinters. The results appear to be entirely satisfactory for qualitative evaluation.—G.F.

High-energy wet scrubbers can satisfactorily clean blast-furnace top gas W. W. Campbell and R. W. Fullerton (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 329–334) The authors summarize data obtained to date in the operation of high-energy wet scrubbers of the orifice or venturi types, at the Clairton, Lorain, and Duquesne Works of United States Steel Corp. It is shown that the scrubbers can clean blast-furnace gas to a degree that is acceptable for gas-fired auxiliaries. The relative merits of the orifice and venturi types are considered.

The economic utilization of slimes obtained in purifying blast-furnace gases K. Niemiec (*Hutn. Listy*, 1960, **15,** (2), 96–101) The slimes were found to be magnetic, and magnetic separators were used to give concentrates with 62%Fe and only 5%SiO_2, the output per concentrator being about 10 t/h. Possibilities of increasing the density of such concentrates, and problems of transportation are considered.—P.F.

Dephosphorization of Indian pig iron (*Nat. Met. Lab. Ann. Rep.*, 1958–59, 8) With high-Si pig iron, dephosphorization in the basic lined cupola was limited and better results were obtained when low-Si pig iron made by oxygen lancing of the high silicon pig iron in the direct arc furnace was charged. With the cold blast, up to 40% phosphorus removal was obtained.

Iron refined in the mixer with oxygen and water G. F. Chub (*Stal'*, 1960, (5), 426) A note from Zaporozhstal'. Blowing for 30 min removed 18%Cr and 11%Mn. Two periods of 1 h removed 25–30%Si and 14–25%Mn. Blowing for 5 h with charging of 25 t scale and 7 t lime with 440 t iron reduced Si 60% and Mn 38%, without change of S.

Metal saving in the transport of the liquid iron S. Popa (*Met. si Constr. Masini*, 1959, (5), 393–397) The optimum capacity of carrier ladles is considered and statistics are given for all parameters such as the number of ladles required per production unit of 100000 t, number of castings, per ladle loss of metal, consumption of refractories, etc.

Pilot plants (*Nat. Met. Lab. Ann. Rep.*, 1958–59, 28–33) A description is given of the design and operation of the low-shaft furnace pilot plant set up by the laboratory at Jamshedpur and the raw materials used. A flow sheet and layout diagram of the plant are given.

Comprehensive tests on self-fluxing briquettes for smelting in the low shaft furnace (*Nat. Met. Lab. Ann. Rep.*, 1958–59, 34) The room temp. crushing strength of the briquettes was found to increase with higher forming pressure. In the SHN testing machine the results have shown that the briquettes having the higher forming loads tend to crumble under a higher applied pressure and a higher temp. The briquettes contained 44% iron ore, 34% non-coking coal, and 6% pitch.

The low-shaft furnace at Barbil (India) W. Wenzel (*Stahl Eisen*, 1960, **80,** May 26, 763) A short description of 36-m^3 capacity low-shaft furnace with a hearth diameter of 2·5 m that has been recently erected in India. The output of the furnace is about 80–85 t/day, the coke consumption about 100 kg higher than in a conventional blast-furnace. A coking plant is being erected for producing suitable coke for the low-shaft furnace.—T.G.

Investigations on the operation of a large circular low-shaft furnace K. Säuberlich and G. Brückner (*Neue Hütte*, 1960, **5,** March, 137–143) A blast-furnace of VEB Maschütte was operated experimentally as a low-shaft furnace, with a charge level of 6·5 m. Best results were obtained by grading the burden to 10–40 mm, and a satisfactory basic Bessemer pig iron was made. Experiments were also carried out using brown coal high temp. coke with the same size range.

The present position of the non-traditional processes for the reduction of iron ore A. Scortecci and A. Palazzi (*Met. Ital.*, 1960, **52,** May, 214–228) As a result of the shortage of coking coal in Europe steps have been taken to economize by using richer ores, concentrating lean ores and sintering. Research into non-traditional reduction processes has been spurred by this situation. Progress with various methods is described. The conclusion is that the low-shaft furnace is at present the only method capable of use on an industrial scale. Methods using electric energy and the sponge iron processes are too costly compared with the blast-furnace in spite of their technological success. A wide field for research still exists.

PROPERTIES, TREATMENT AND USE OF SLAGS

Utilization of slag obtained in processing stainless steel scrap in the electric arc furnace, using oxygen J. Chovjka (*Hutnik*, 1960, **10,** (1), 2–5) Chromium can be recovered for further use in alloy steel by adding Cr-bearing slags to reducing slags of melts in alloy steel production. In this way the Cr is utilized without a deterioration in the steel. The method of salvaging Cr is cheap to introduce. In the author's experiments, carried out in a 2-t basic electric arc-furnace, Cr ores containing 37%Cr_2O_3, and hence 25%Cr, were successfully utilized. The experiments and results obtained are discussed.—P.F.

The importance of physico-chemical investigations to the development of granulated slag technology T. Rogozinski (*Hutnik*, 1960, **27,** (3), 109–114) In order to increase the use of granulated slag in the cement industry, investigations into the structure of these slags have been carried out. A hypothesis has been advanced that the binding properties of the slag depend on the size of ions and increases as ion size decreases. In particular the properties of Al ions were discussed since the best slags have an Al_2O_3 content of above 13%. The apparatus used in the investigations is described.

Activity of the components of the CaO–Al_2O_3–SiO_2 fused systems I. T. Sryvalin and O. A. Esin (*Izvest. VUZ Chern. Met.*, 1959, (8), 9–16) An expression is proposed for calculating the activity of the components of this fused system. The equation is obtained by introducing certain corrections in the equation for regular solutions. The values of the activities of CaO and SiO_2 calculated by means of this equation agree well with experimental data.

Visualizing crystal growth in melts and measuring liquidus temperature to 2300°C with microscope equipment G. Cavalier (*Amer. Ceram. Soc. Bull.*, 1960, **39,** March, 142–144) The melt held by capillary effect on a hairpin-shaped Pt or Mo wire is heated to the liquid state by ac current flowing through the wire, observation being possible by illuminating the sample under a powerful lamp. A tiny thermocouple is introduced into the melt to measure temp. when crystals appear round the junction. Blast-furnace slags were investigated. Accuracy = ±3°C at 1600°C.—S.H.-S.

Electrical conductivity and magnetic properties of melts of the ternary oxides FeO–SiO_2–CaO (ZnO.Al_2O_3) V. P. Schastlivyi and D. M. Chizhikov (*Izvest. AN USSR, OTN, Metallurg. i Topl.*, 1959, (6), 16–20) The measurement of the specific electrical conductivity was carried out from volt-ampere readings with two 3·5 mm dia. W electrodes between which a Pt–Pt–Rh thermocouple was inserted, in corundum crucibles in an atmosphere of N_2, heated in a Silit resistance furnace. The electrical conductivity of this liquid system obeys an exponential law. The activation energy does not exceed 18 cal. The magnetic permeability is practically independent of temp. between 1100–1450°.

DIRECT PROCESSES

New electric direct reduction process producing steel or pig iron A. De Sy (*AIMME, Blast Furn. Comm. Proc.*, 1959, **18,** 316–327) The author describes a new direct reduction process which is operated in a double-hearth low-frequency induction furnace. It requires a special high-grade iron ore and a fairly pure carbon as reductant, and produces either a rimming steel or a carbon-rich iron. The principle of the process is described and experimental results are discussed. The practic-

ability of the process is also considered.—G.F.

Electric steel by direct reduction P. Nigri (*Tek. Tidsk.*, 1959, **89,** Oct. 9, 973–976) A description is given of the Tysland-Hole low-shaft furnaces, and the Strategic-Udy, Lubatti, and von Roll processes.

PRODUCTION OF STEEL

What's in a name? T. S. Kilpatrick (*Rail. Steel Topics*, 1960, **5,** (3), 15–22) The choice of a steelmaking process is largely dictated by factors the steelmaker does not control. It is suggested that engineers should be concerned with the physical results of manufacturing practice rather than specifying the processes which produce them. The acid Bessemer process is described to show that although different processes produce steels with different attributes these are well understood and capable of being demonstrated and tested.

Acme Steel Company (*Iron Steel Eng.*, 1960, **37,** March, A1–A44) A review of the development and present operations and capacity of the Company particularly at the main works at Riverdale, Illinois.

Davy-United's contribution to Durgapur 1st stage (*Anvil*, 1960, **5,** Spring, 18–27) A detailed description is given with photos and layout diagrams of the design and operation of the completed stage of the works. The progress of stages II and III is also dealt with.

Steelmelting shops S. K. Kar and T. V. S. Ratnam (*TISCO*, 1960, **7,** Jan., 1–68) The various steelmelting shops of the Tata Iron and Steel Co., are described in detail, and the developments in the 'new' No.3 steelmelting shop, and the planning behind these developments, are minutely described.

Some major problems in steel plant engineering for growth at Stelco L. H. Chater (*Eng. J.*, 1960, **43,** May, 41–49) Materials handling at the Hamilton Works (Steel Co. of Canada) including iron and slag, and some constructional problems are discussed.

Pilot plants in pyrometallurgy N. Booth (*NML Techn. J.*, 1960, **2,** Feb., 47–51) Advantages and disadvantages of pilot plants in pyrometallurgical processes are described and size, location, and design are discussed. Thermal and chemical considerations are raised and special features, from processes in which British Oxygen Research and Development have worked, are described.—S.H.-S.

How steelmakers can use oxygen safely R. E. Armstrong (*Steel*, 1950, **146,** Feb. 29, 88–90) Advice is given on the safety precautions which are desirable when using oxygen in steelworks. Oxygen pipes should be free of burrs and scale and equipped with filters; they can be cleaned with caustic solution. Oxygen lines should be marked and protected from oil and mechanical damage. Lancing equipment must be carefully used and maintained.—D.L.C.P.

Oxygen production and distribution centre at the Oxylor Works (*Techn. Mod.*, 1960, **52,** April, 194–196) The article describes the main features of the Oxylor oxygen production centre just completed near Thionville by Air Liquide to supply the Lorraine steelworks. The supply network extends over a 25-mile radius snd feeds ten works. Extensions to Nancy in the South and Longwy and Luxembourg in the North are planned. Supply lines enable 2120×10^6 ft^3 of oxygen to be handled.

The use of oxygen in steelmaking T. B. King (*AIMME, Proc. OH Comm.*, 1959, **42,** 250–262) The author discusses the functions of oxygen in steelmaking processes, and considers the elementary thermo-chemistry and kinetics of such processes. He deals separately with O_2 enrichment of combustion air and with the use of O_2 for accelerating the refining reactions. Attention is also given to P elimination at high carbon levels and the production of low-nitrogen steel and very low-carbon steel.—G.F.

Low-sulphur steelmaking proves out on tonnage basis H. W. McQuaid (*Iron Age*, 1960, **185,** Feb. 18, 114–116) It is reported that the 'rotary reactor' patented by N. Goss, has proved successful under production conditions. In the process, liquid metal is spun in a rotating reactor with caustic soda and, if required, iron oxide and oxygen; the slag and hot metal are then separated. In this way, S, Si, C, and P contents of the metal are considerably reduced. Use of the process would give more freedom in blast-furnace operation. The product would give economies in or extend the applications of OH furnaces, converters, or electric furnaces. A 60-t unit has been in operation.—D.L.C.P.

Scrap handling programmes. How to plan them for profits (*Austral. Mach. Prodn. Eng.*, 1960, **13,** Jan., 17–20) Factors involved in the design and installation of a scrap handling system are reviewed in general terms. The system installed at GEC Louisville is briefly described.

Reclamation and processing of scrap from steelmaking slag and Refuse-Heckett process G. G. Spandl and A. M. Sundaram (*TISCO*, 1960, **7,** Jan., 69–76) The Heckett process for reclamation and processing of scrap is described and illustrated with photographs of the plant and equipment in operation at the Tata works in Jamshedpur.

Determination of the characteristics of a press for scrap pipe and tube R. Ramiondi (*Met. Ital.*, 1960, **52,** March, 73–77) This article describes a study made to establish the need for installing a press for baling tube scrap and crop ends for charging into electric furnaces. The calculation of the constants which appear in the formulae used to determine the performance of the press is given. The required characteristics of the press obtained by means of this study are tabulated.

Keep lime flux moisture free Timken Roller Bearing Co. (*Iron Age*, 1960, **185,** Feb. 4, 89) A system using waterproof containers for lime in an electric melting shop is described. Lime is stored in the portable metal containers, and emptied into the charging boxes by fork-lift truck. Objects are the prevention of water absorption by the lime and of the production of H_2 in the quality steel, saving of space and handling, and reduction of dust nuisance.

Some applications of computer control in the iron and steel industry T. M. Stout and S. M. Roberts (*Iron Steel Eng.*, 1960, **37,** March, 101–108) Types of computer control systems are described, typical cost figures given, and some applications to iron and steel manufacture suggested (17 refs).

On the production of soft magnetic steel in the oxygen converter L. Kodrle (*Hutn. Listy*, 1960, **15,** (1), 13–20) Full-scale experiments have shown that the low-carbon steels produced in the oxygen-blown converter are at least equivalent to corresponding OH steels. Particularly low gas contents are attainable if the rimming steel is subjected to a vacuum; the carbon in the melt simultaneously contributes to the deoxidation of the metal.—P.F.

Steel making by L.D. process M. Ganapati (*Iron Steel Rev.*, 1960, **3,** Jan., 19–23) The origin and development of the LD oxygen steelmaking process are described, and the advantages of the method outlined. The design capacity, and plant of the Rourkela Steel Project are then discussed.

On the lining of the oxygen converter S. Machara and T. Kohtani (*Tetsu to Hagane*, 1959, **45,** Sept., 957–959) Details are given of an examination of used brickwork, and of the schedule for relining.—K.E.J.

Study on a permanent wall type furnace with forced cooling. IV. Operation of the water-cooled converter. 2 R. Sato, O. Seya, and N. Tabata (*Tetsu to Hagane*, 1959, **45,** Nov., 1255–1260) Using a side-blown converter process, low-P steel could be made with high C contents. The process largely eliminates Cr during refining. Refractory materials for patching must be chosen to combat erosion by dynamic streams of slag, metal, and blast.

Melting trials on converter skulls J. Sophron (*Centre Doc. Sidér. Circ.*, 1959, **16,** (4), 847–860) A series of experiments is described in which nose skulls are prevented or removed by the use of an oxygen lance.

Manufacture of mild killed basic Bessemer steel M. le Moyne (*Centre Doc. Sidér. Circ.*, 1959, **16,** (4), 861–891) A detailed account is given of the analysis, killing technique, casting and chemical and physical properties of the low-C killed steel made at Lorraine-Escaut.

Engineering aspects of a modern open hearth steel plant J. A. Kilby and J. C. Gordon (*Iron Steel*, 1960, **33,** May 25 (special issue), 276–281) A West of Scotland ISI paper.

On the reconstruction of a 20-ton basic open hearth furnace I. Hisano, H. Kato, and H. Tokunaga (*Tetsu to Hagane*, 1959, **45,** Sept., 942–944).—K.E.J.

X-Ray studies on the refractories used in the roof of a basic open hearth furnace T. Ao and T. Oyama (*Tetsu to Hagane*, 1959, **45,** Sept., 946–948) Data are given for cristobalite, matrix material, and spinel solid solns. from different zones in used bricks (13 refs).—K.E.J.

Fundamental research on combustion in open hearth furnaces. II. Influence of atomised steam upon combustibility in open hearth furnaces Y. Okada, K. Nakamachi, T. Kinoshita, and M. Yamamoto (*Tetsu to Hagane*, 1959, **45,** Sept., 952–955) A statistical analysis relates the steam used to flame length, draught, oxygen in waste gas, etc., and operating results show the effect of steam on flame length and combustion in the various stages.

On the study of iron balance in open hearth furnaces T. Kai, H. Watanabe, and S. Fuzita (*Tetsu to Hagane*, 1959, **45,** Sept., 968–970) The increase in steel yield with pig ratio (in the range 55–70%) is shown, and a complete iron balance is constructed.—K.E.J.

A new technique for heating open-hearth furnaces with cold gas and heating oil. Part I W. Schmitt (*Stahl Eisen*, 1960, **80,** April 28, 616–619) The requirements on the flame of an OH furnace during melting down and during refining are described and the principles of the new heating explained: During melting down a short, wide, hot, non-luminous coke-oevn gas flame is provided by the new burner, during refining a long, highly-luminous flame covering the whole of the bath is used. This flame is obtained by burning heating oil alone or in combination with cold gas. **Part II** O. Dürmann (619–621) Results on 75- and 90-t OH furnaces with the new cold-gas/oil heating method are reported. Whereas the use of oil alone resulted in a high wear and short life of roof and lining, the new combination of cold gas and heating oil, described in Part I, brought about—after slight modification of the furnace—a substantial increase (up to twice) in the life of the roof and lining, a higher output (up to 15%), and a higher calorific efficiency. The new burner is marketed by Demag.—T.G.

Progressive instrumentation: One key to a better future D. E. Brophy (*Blast Furn. Steel*, 1960, **48,** Jan., 45–50) The author reviews the impact of instrumentation on the operation of OH furnaces.—T.G.

On the manufacture of fine grained austenitic steels in the open-hearth furnace T. Myslivec and S. Kubêna (*Hutn. Listy*, 1959, **14,** (12), 1025–1032) The joint addition of Al and Ti as de-oxidizers and grain refiners was studied in medium carbon steels and low alloy Cr–Mn steels. The addition of both elements, rather than Al alone, improves the low-temp. notch impact-strength and the stability of the small grain size attained, particularly at elevated temp. Grain sizes measured on samples taken from the melt are not a reliable indication of the final grain size attained in the finished, rolled product.—P.F.

On the desiliconisation of hot metal with oxygen blowing M. Maeda, S. Tsuzuki, and S. Hayashi (*Tetsu to Hagane*, 1959, **45,** Sept., 883–885) Details are given of the reduction of C, Mn, and Si, and of the value of the hot metal in the OH furnace. 99% O_2 is more effective than 70%.—K.E.J.

Kinetics of decarburisation in molten steel. II. The rate of decarburisation in the basic open hearth T. Fujii (*Tetsu to Hagane*, 1959, **45,** Oct., 1135–1139) Results from a 100-t furnace are given. For [C] < 0·30%, the rate of decarburization decreased in proportion to the carbon content in the molten steel (equation given). No relationship with Δ[O] was found. The rate of decarburization when the oxygen lance was used was large compared with the ore method. **III. Effect of carbon and gaseous oxygen in molten steel on the rate of decarburisation** (Nov., 1248–1254) Laboratory experiments show that, in argon, the decarburization rate is constant (reaction with the MgO crucible). With constant oxygen concn. in the gas, the rate was independent of C concn.

above 0·15%; below this it fell rapidly. The rate was constant for increasing oxygen concn. above 10%, as a result of surface-film activity. **IV. Effect of temperature, depth of molten steel, free surface, and rotation on the rate of decarburisation** (Sept., 932–935) Data are given from a high-frequency induction furnace with atm. of A + 8% or 16% O_2.—K.E.J.

Effects of various factors on the rate of desulphurisation of molten pig iron with solid lime T. Yagi and Y. Ono (*Tetsu to Hagane*, 1959, **45**, Nov., 1236–1241) Desulphurization hardly proceeds in an oxidizing atm. containing CO_2. The reaction of molten Fe–C–S in neutral or reducing atm. is initially first-order with respect to S content. The rate in town's gas is thrice that in CO or A. Lime should be pulverized to give a large contact area. The rate increases with increasing temp. and Si content.—K.E.J.

Use of oxygen lances and basic brick in open hearth furnace roofs H. A. Parker and P. Schane jun. (*Blast Furn. Steel*, 1960, **48**, Jan., 68–74, 78–80) Two technological improvements are described that have materially contributed to increasing the steel-producing capacity of OH furnaces, they are: the use of oxygen lances suspended from the roof for reducing heat time and the development of a basic roof construction that is economical and less sensitive to the wide variations in temp.

Change of hydrogen content in the bath during melting in the open-hearth furnace T. Myslivec (*Hutn. Listy*, 1959, **14**, (9), 749–753) According to tests on 50-t open-hearth furnaces, H_2 loss is promoted in the melting-down period by the high FeO content of the slag. During the boil H_2 content rises, and in the final stages of the melt a further rise is observed due to high bath temp. and preliminary deoxidation in the furnace, and when the metal was held after preliminary deoxidation (11 refs).

Effect of basic open hearth furnace conditions on internal defects of killed steel tubes Y. Kitamura, A. Cho, and S. Tomita (*Tetsu to Hagane*, 1959, **45**, Sept., 972–974) Roof life is correlated with length of internal defects, total Fe, bath temp., and ladle analysis, and length of defects is also correlated with total Fe.

Absorption of metal by the hearth of a steel furnace V. M. Zamoruev (*Stal'*, 1960, (3), 223–225) Alloy-forming elements tend to absorb in the hearth and to be extracted by subsequent heats. W in magnesite is especially studied using ^{185}W. Losses of 3–7·6%W are sustained by absorption. The use of fine-grained magnesite to cut down hearth porosity reduces this.

International flame research on open hearth firing F. W. Luerssen (*AIMME, Proc. OH Comm.*, 1959, **42**, 228–247) The author reviews the work of the International Flame Research Foundation at IJmuiden, drawing attention to the studies that are particularly relevant to OH furnace operation. He discusses the development of instruments for the measurement of combustion variables; studies of the radiation characteristics of fuels and other combustion variables such as type and quantity of atomizing agent, air preheat, degree of mixing, and jet momentum; and investigations into burner design.—G.F.

Fuel practice at Fontana H. I. Riegel (*AIMME, Proc. OH Comm.*, 1959, **42**, 219–226) Data are presented showing the gradual improvement in OH furnace performance at the Fontana Works of Kaiser Steel Corp. since the war. The author discusses this improvement and describes the changes which have been made in furnace design, firing practice, and operation.—G.F.

Problems involved with high percentage hot metal charge R. Duffett (*AIMME, Proc. OH Comm.*, 1959, **42**, 52–55) The author presents data obtained in the two OH shops of Great Lakes Steel Corp. on charges ranging from 45% to 68% hot metal, and discusses some of the problems which can arise with high hot-metal charges. The general indication is that hot-metal percentages in the low sixties offer the most productive charge.—G.F.

Investigation with the aid of radioisotopes on the wear of refractory runner bricks and its effect on the purity of structural alloy steel H. Treppschuh, E. Pachaly, K. Sauerwein, and R. Schröter (*Stahl Eisen*, 1960, **80**, June 23, 878–882) The runner system was labelled with the ^{45}Ca isotopes and the various part of the runner system were studied separately. The results show that the wear products from the refractory bricks were almost completely collected in the sinkhead and that in fact no contamination of the steel occurred from the refractory bricks.—T.G.

On the corrosion of low-flux silica roof bricks S. Kienow (*Stahl Eisen*, 1960, **80**, June 23, 882–888) The behaviour of the roof bricks of an OH furnace was studied. The author found that after 659 charges of the furnace the bricks showed a structure that could be divided into four zones. Entering of melt into the bricks had a densifying effect on the bricks, any iron oxide present was pushed towards the centre of the brick. Recrystallization of the bricks caused shrinkage and the pitting observed is also attributed to the effect of recrystallization.—T.G.

Uses of substitutes for pig iron L. W. Krum (*AIMME, Proc. OH Comm.*, 1959, **42**, 57–59) The author summarizes replies to a questionnaire designed to assess the desirability of using carbon substitutes for pig iron in acid OH practice. It is shown that substitutes are used by most operators for all regular grades of acid steel, and no major drawbacks are noted.

Use of iron substitutes H. W. Potter (*AIMME, Proc. OH Comm.*, 1959, **42**, 312–314) Relative scrap prices in the Philadelphia area indicate the desirability of reducing the pig iron charge in cold metal OH practice to a minimum. The author considers the economics of seven different charges in which the iron is progressively reduced and replaced largely by No.2 scrap, turnings, coal, and coke. Allowance is made for influencing factors, such as variations in Mn and Si contents at melt, different charging times, and the foaming hazard. The use of injected graphite to remedy soft melts is briefly discussed.—G.F.

High flush practice at Algoma T. R. Conway (*AIMME, Proc. OH Comm.*, 1959, **42**, 55–56) Typical OH charges and the charging pattern at Algoma Steel Corp. are given, and the problems encountered in high-flush practice, with up to 80% hot metal, are outlined.—G.F.

The state of development of all-basic open hearth furnaces in the German Democratic republic F. Rabes and L. Ullsperger (*Neue Hütte*, 1960, **5**, April, 218–232) A review of practice outside Germany is followed by an account of East German all-basic practice, including a description of an improved roof, and details of construction above and below platform level. Economic aspects are considered.

Variations in the pressure in the vertical gas uptakes of O.H. furnaces working with compressed air V. S. Kocho and V. I. Grankovskii (*Izvest. VUZ Chern. Met.*, 1959, (6), 93–97) Fuel consumption of large OH furnaces on mixed gases may reach 8000–12000 t/m³/h and this increases the pressure in the gas uptakes. Figures quoted in support of this fact give the variations of the pressure in the gas uptake of a 250-t furnace at the Alychevsk Works in relation to gas consumption. The temp. of the gases also affects the pressure. If an automatic control of the consumption of compressed air is installed, regulating the heat load, it is possible to select such a cross-section for the air tuyeres that the pressure in the uptake will only slightly exceed atmospheric pressure.

Distribution of the iron oxides in the height of the slag layer in the bath of the 500 t O.H. furnace E. V. Chelishchev (*Izvest. VUZ Chern. Met.*, 1959, (6), 23–30) The distribution is not uniform. The greatest concentration of ferric oxide is found at the gas–slag interface, of ferrous oxide, between the metal and slag. The transfer of slag volumes from the zone of one phase to that of the other causes the reduction of the higher oxides at the metal slag interface and the oxidation of the lower oxides at the boundary between slag and gas. The degree of the non-uniformity of the iron oxides depends on the degree of mixing taking place in the bath, which is itself conditioned by the rate of combustion of carbon.

Increased open hearth production with the use of oxygen T. A. Thompson (*Iron Steel Eng.*, 1960, **37**, March, 121–124) Weirton's experience using O_2 roof lances is reviewed, and a description is given of the record tap-to-tap time of 5 h 35 min, corresponding to a steel production of over 100 t/h.

Effect of combustion oxygen on open hearth production J. L. Evans (*AIMME, Proc. OH Comm.*, 1959, **42**, 268–273) O_2 enrichment of combustion air is routine practice at Wheeling Steel Corp. The author describes the practice and discusses its effect on operating variables such as charging time, fuel requirement, and production rate.—G.F.

Oxygen injection for decarburization G. S. Baldwin (*AIMME, Proc. OH Comm.*, 1959, **42**, 318–319) Experience in oxygen lancing in the OH furnaces at Standard Steel Works of Baldwin-Lima-Hamilton Corp. is briefly reviewed. The practice has resulted in shorter heat times coupled with improved quality.

Basic open hearth checkers W. W. Bergmann (*AIMME, Proc. OH Comm.*, 1959, **42**, 197–199) Basic-capped checkers at Armco Steel Corp. have reduced replacement costs, heat time, labour costs, and furnace downtime. Operating data are briefly reviewed.—G.F.

Open hearth roof repair with fuel on A. J. Voss (*AIMME, Proc. OH Comm.*, 1959, **42**, 188–193) Details are given of the method adopted at Inland Steel Co. for OH roof repair with fuel on. Use is made of a metal-cased basic brick provided with tabs which overlap the rib brick.—G.F.

Gun-emplacement silica ganister in the open hearth J. V. Grothaus (*AIMME, Proc. OH Comm.*, 1959, **42**, 179–185) Gun emplacement of clay-ganister mixes is used at Great Lakes Steel Corp. for transfer ladles, hot-metal spouts, OH bifurcated launders, and several spot applications. The advantage of gunning compared with ramming are discussed.—G.F.

Problems encountered with other sections of basic roof furnaces W. J. Scharfenaker (*AIMME, Proc. OH Comm.*, 1959, **42**, 151–158) The author stresses the need for minimum repair and maintenance time and maximum availability of the whole furnace unit so that the full value of basic roofs may be realized. He deals particularly with front and back walls, furnace ends, and the slag pockets and checkers.—G.F.

The basic roof at Fairless Works H. A. Parker (*AIMME, Proc. OH Comm.*, 1959, **42**, 135–144) The author describes the experimental programme started in 1957 at Fairless Works of United States Steel Corp. to determine whether basic roofs could be used economically with O_2 injection by roof lances. The roof design is described and illustrated, and some problems still remaining are discussed. Operating data from five campaigns are summarized, showing the benefits of low roof costs, increased production, and less downtime.

FOUNDRY PRACTICE

Acid cupola melting for ductile iron H. E. Henderson (*Mod. Castings*, 1959, **36**, Nov., 661–668) Cupola practice at the Lynchburg Foundry Co., Lynchburg, is described with reference to control of composition. Limited tests using a basic cupola were unsuccessful in producing iron of the correct carbon and sulphur contents.—A.D.H.

Basic cupola melting of ductile iron J. T. Williams (*Mod. Castings*, 1959, **36**, Nov., 669–670) Basic cupola practice at Grede Foundries Inc., is described and compared with acid practice.—A.D.H.

Practical tips on cupola control S. F. Carter (*Foundry*, 1960, **88**, Feb., 88–95) The variables in cupola operation are considered in turn: the theoretical and practical effects of each are shown and advice is given on their control. The following are dealt with: coke composition, size and ratio, air volume, charge distribution, bed preparation and height, scrap and stone size and shape, slag control, linings, tuyeres, and dams. Graphs and tables are given to illustrate the effects.—D.L.C.P.

Cupola blowers with rotating impellers M. Gregoraszczuk and L. Zurawski (*Prz. Odlew.*, 1960, **10**, (2), 47–53) In the article the principles of design and work of blowers with Root and Jaeger type rotor impellers are discussed.

Formulas for calculating the actual output of the blowers and the power requirement are derived. Features which will be of most interest to users are given and comparison of this type of blower with a centrifugal blower is made, considering mainly the coordination with a cupola.

Natural gas in the foundry. Types of burner used (*Fonderie*, 1960, Feb., 70–73) The characteristics of natural gas are discussed in relation to various types of burners used in foundries. Possible applications of natural gas are indicated. Town gas and natural gases are compared.—R.P.

The use of methane gas for melting cast iron in cupolas C. Stefanescu, T. Steflea, and S. Teodorescu (*Met. Constr. Masini*, 1959, (8), 671–681) The use of methane in converted cupolas in additional gas burners has reduced coke consumption by 7·5% and will probably reduce it further up to 15%, with the improvements now being tested. CH_4 consumption is at the rate of 40 m³/t of iron, representing a reduction of 26% in the calorific consumption as compared with coke-fired cupolas. It has reduced lime consumption by 50%, increased the yield from 1·5 to 2·3 t/h. Temp. at pouring was 1360° (by optical pyrometer without correction). In the iron it has reduced the carbon content by 2–3%, and the S content by $1 \times 2{\cdot}5\%$

A new mechanized cupola charging system S. Komorowski (*Prz. Odlew.*, 1960, **10,** (2), 37–42) The article consists of an illustrated description of a mechanized cupola, charge preparation and charging system. The system described was built by the Constructional Engineering Co. Ltd., Birmingham, for the Mill Top Foundry Co. Ltd, Birmingham.

The M.B.C. Cupola and its application M. M. Foulon (*Prz. Odlew.*, 1960, **10,** (2), 42–47) The distinctive features of the MBC cupola are a pre-heated air intake and a strongly basic slag under reducing conditions. As a result it is possible to use a charge containing a large proportion of scrap steel. A very considerable saving is obtained by limiting loss through oxidation. As the body of the cupola is lined with a carbon preparation and the shell in the melting zone is not lined with any refractory material, repairs are infrequent. This cupola can be used extensively in the foundry and as a melting unit in co-ordination with a steel-making furnace.

Prospects of new developments in the cast iron furnace process P. Murza-Mucha (*Prz. Techn.*, 1960, **81,** (8), 4–6) The author discusses the cupola with a pre-heated air intake considering in particular the MBC furnace. This type of furnace is not yet used on an industrial scale in Poland. An important advantage of this furnace is that coke briquets made from ordinary coal can be used. The production of these briquets in Poland is in an advanced stage.

Gas firing cuts foundry costs S. J. Michaels (*Foundry*, 1960, **88,** Feb., 128–132) Results from a foundry which has found natural-gas heating preferable to oil firing are described. Gas-fired reverberatory furnaces are used for melting metal for ductile iron castings. Heat-treatment furnaces and other applications have also been converted to gas heating. Although straight fuel costs are higher, total costs are lower due to savings in maintenance of refractories and burners, alloying materials, spoilage, production time lost and labour costs.

Cupola meter G. Salter Ltd (*Inst. Practice*, 1960, **14,** Feb., 149) A blast meter installed on a 22 cwt cupola and the improvements effected in operation are indicated.

Pig iron from reclaimed scrap steel Ferranti Ltd (*Engineer*, 1959, **208,** Dec. 11, 786) A 3-t electric-arc furnace installation is briefly described. Croppings from transformer laminations, other mild steel, foundry scrap, and a recarburizing compound are used.

High-speed melting J. G. Winget (*Mod. Castings*, 1959, **36,** Sept., 44–46) The development by Roda Pump Co. of Oklahoma of a rapid-melting gas/oil fired furnace is described. The furnace has a capacity of about 600 lb, works under a slightly reducing atmosphere, and is used for ferrous and non-ferrous work.

Comparative study of medium- and industrial frequency induction furnaces for ferrous metals A. R. Baffrey (*Fond. Belge*, 1959, **29,** (11–12), 281–285) The main features of the two types are described and compared with examples. The medium frequency types (500–8000 Hz) are recommended for furnaces of less than 1·5 t capacity, and large furnaces started frequently with cold charges. Mains frequency furnaces should be used for larger furnaces.

Indirect arc electric furnace production of ductile iron L. Miller (*Mod. Castings*, 1959, **36,** Nov., 653–660) The operation of a 20-t rocking arc furnace for the production of SG iron and the use of a plunger ladle are described.

Furnace power cost cut by control through television L. Dirrig, J. Robertson, and R. A. Aschenbrener (*Foundry*, 1960, **88,** Jan., 142–146) The power demand of a 3-t, top-charging melting furnace has been considerably reduced and substantial saving in money achieved by televizing the power company demand meter readings to the furnace operators. The scheme is described in detail and the economic justification for the elevision equipment discussed.

The shaking ladle—a new metallurgical aid, with special reference to the manufacture of high-grade cast iron S. Eketorp and B. Kalling (*Giesserei*, 1959, **46,** Nov. 5, 905–912) The design and operation of the ladle, and several examples of its use in desulphurization, recarburizing and alloying iron, and of refining to remove Si and V, are described.

Solidification of cast iron C. K. Donoho (*Mod. Castings*, 1960, **37,** April, 169–171) The carbon equivalent is explained, and various formulae for its calculation are presented. The solidification of hypo- and hyper-eutectic white, grey, and malleable irons is then considered.

The effect of the manganese content on the shrinkage of grey cast iron L. Sofroni and C. Cosneanu (*Rev. Roumaine Met.*, 1959, **4,** (2), 145–156) [In German] With Mn contents up to 5% in normal grey cast iron, with small P percentages, the initial expansion increases only very little, but with 7–9%Mn it increases considerably (more than twofold) followed by sudden shrinkage. With higher P contents no appreciable increase in the initial expansion can be observed, but the sudden shrinkage occurs at smaller Mn contents. Beginning with 8–9%Mn the amount of carbide increases considerably, resulting in an increase in hardness up to 450 H_B.

On the properties of nodular cast iron T. Momose and T. Izu (*Imono*, 1959, **31,** Aug., 707–711) Effects of Ti and Sb are investigated. Neither has much inhibiting action separate but together (0·1%Ti, 0·003%Sb) spheroidization fell to 40%.

Fundamental studies on spheroidal graphite cast iron. Influence of zirconium on formation of spheroidal graphite S. Morita, K. Kurai, and M. Tsuji (*Tetsu to Hagane*, 1959, **45,** Sept., 907–909) The formation of perfectly spheroidal graphite is shown to depend on the relationship between the residual contents of Zn and Mg.

Nodular iron gears stand up to overloads and rough use B. Hopper (*Iron Age*, 1960, **185,** Jan. 14, 74–75) Evidence is presented which suggests that nodular iron gears are as good as steel gears from the standpoint of pitting, scoring, and beam strength.—D.L.C.P.

Hydrodynamic, fundamental and practical study of finger-gate feeding systems M. Jeancolas, G. C. de Lara, and H. Hanf (*Fonderie*, 1960, March, 89–110) The theory of gating and running is discussed in detail and related to practical aspects (22 refs).—R.P.

New method of manufacture of flanged steel castings J. Přibyl (*Giessereitechn.*, 1959, **5,** Sept., 253–257) The method described is applied to the manufacture of three- and four-flanged castings, in which cooling conditions are so devised that one riser is sufficient. The method is suitable for series production, and it is thought possible that it could be applied to the manufacture of nodular iron castings (10 refs).

Capacity and output per unit area in green-sand machine-moulding H. Geissler (*Giessereitechn.*, 1959, **5,** Aug., 225–234) Laws governing the determination of capacity are discussed, and methods of calculation are worked out with examples. Melting practice is taken into account in the calculations. Output per unit area of the moulding shop (t/m²) is derived directly from the factors obtained in the determination of capcaity.

Thermal efficiency of drying stoves (*Found. Trade J.*, 1959, **107,** Oct., 387–388) A description of a test for stove efficiency is given, and then the effects of firebox capacity, control of combustion, and temp., stove leakage, chimney draught, and stoving cycle and insulation are considered.

Oil heated core drying ovens Sulzer Bros. (*EFCO J.*, 1960, **1,** Jan., 14–15) Four ovens with a total capacity of 160 t for drying resin-bonded cores are briefly described.

Air-setting cores A. M. Shirley (*Brit. Found.*, 1959, **52,** Nov., 502–507) Air-setting compounds used with a sand-slinger increased core production. Examples of castings produced and the savings obtained are given.

Use of the clay waterglass process in steel casting F. Dubielzig (*Giessereitechn.*, 1959, **5,** Sept., 258–261) This process is compared with other moulding methods, particularly the normal CO_2 process. Its operation is described, with several examples; castings up to at least 1200 kg can be made. Space is saved and substantial economies are possible.

Optimum CO_2 moulding: critical formulations of sodium silicate and sand R. J. Cowles (*Mod. Castings*, 1959, **36,** Sept., 91–114) Topics discussed include laboratory test methods to improve collapsibility by determining the best proportion of silicate binder, improvement of sand properties by the use of additives, and a general discussion on the properties and application of binders (34 refs).

Mould and core inspection can cut cleaning room costs J. W. Gillespie (*Foundry*, 1960, **88,** Feb., 80–83) An example of cost reduction in the foundry is described; it was found that extra care in mould and core preparation was more than paid for by savings in the castings cleaning department.—D.L.C.P.

Development of the technology of a chill mould E. Bakowsky (*Giessereitechn.*, 1959, **5,** Sept., 274–277) Five types of mould are distinguished, and points to consider in the design and workmanship are considered an illustrated by examples. **Technology of the manufacture of casings, pattern no. —, in chill-cast iron** (277–278) Various stages are given in the manufacture of a casing forming one of the four classes of mould described in the above article—i.e. an open mould with a gate.

First practical experiences with the chill mould-casting aggregate GKF 2 M. Dietrich and J. Stengl (*Giessereitechn.*, 1959, **5,** Sept., 262–264) Design and operation of the equipment manufactured by the VEB concern 'Ferdinand Kunert' and tested by VEB Metallguss, Leipzig, are described.

The permanent Chamotte mould A. Bohne (*Giessereitechn.*, 1959, **5,** Sept., 278) This type of mould, used for reasons of economy where the expense of a chill mould was not justified, has been used successfully at Wurzen for casting tubes 2000 mm long, and with external and internal dia. of 315 and 235 mm respectively.

Air-set cores for heavy general-engineering and jobbing castings A. M. Shirley (*Found. Trade J.*, 1960, **108,** March 31, 387–393) The installation, operation, and examples of the work of a sandslinger are described and illustrated.

Investment-casting experience: preformed refractory cores and casting behaviour of Ni-Cr alloy N. Walker (*Found. Trade J.*, 1960, **108,** April 7, 431–436) Core and corebox materials and production and the wax-injection process are described.

Making a 37-ton lathe-headstock casting A. Ross and C. Warburton (*Found. Trade J.*, 1960, **108,** March 17, 325–328) A detailed account.

Primer on shell moulding J. J. Silk (*Mod. Castings*, 1959, **36,** Sept., 48–50, 136) Casting and pattern design, and problems arising in the use of shell-moulding techniques are reviewed.

Factors to consider in producing shell cores R. S. L. Andrews (*Foundry*, 1960, **88,** Feb., 96–100) Characteristics of the two basic types of resin binders and their relationship to types of shell cores and coremaking machines, mix compositions, corebox design, and casting defects are discussed.

Shell moulding developments in the automotive industry H. C. Grant (*Mod. Castings*, 1959, **36**, Nov., 641–652) Details of shell moulding practice in three foundries of the Ford Motor Co. Michigan are described and the advantages of the process are summarized.

Investment shell process offers production advantages R. H. Herrmann (*Foundry*, 1959, **87**, Dec., 84–87) The lost-wax investment casting process is briefly described and its merits and advantages stressed, at the same time its limitations are pointed out clearly.

Introduction of tentative hot shell deformation test R. J. Cowles (*Mod. Castings*, 1959, **36**, Sept., 51–54) A report of the A.F.S. Sand Div., Shell Mould and Core Committee 8–N. A new method is described, and results are analysed in which heat is applied rapidly to one side of shell mould specimens which are loaded transversely at the same time. Information obtained has been used in developing tentative standard equipment and methods for controlling hot deformation in production.

Surface defects on shell-moulded castings J. A. Behring and R. W. Heine (*Mod. Castings*, 1960, **37**, Feb., 58–66) The causes, appearance, and cures for the orange peel defect, surface shrinkage, metal penetration, and carburized surface in castings are briefly reviewed, and then work carried out to determine the causes of the 'shell-mould' (or gas-pit) defect is described. The defect was found to be a surface accumulation of oxides and associated gas, due to the oxygen in the phenol-formaldehyde shell resin. The effects of casting conditions on the occurrence of the defect have been investigated, and methods of minimizing the defect are discussed.

Ramming, superheat and alloys (type of metal) effects on metal penetration G. J. Vingas (*Mod. Casting*, 1959, **36**, Nov., 671–675) Factors influencing the penetration of metal into small masses of sand surrounded by a relatively large mass of molten metal are given. Tests showed that penetration of steel was unaffected by ramming density and superheat but that both variables affected the penetration of cast iron.—A.D.H.

Shrinkage porosity in a grey iron casting (*Foseco Found. Practice*, 1959, Dec., (138), 1–3) An account of a water pump component in which porosity persisted. The risers were not functioning properly, and composition and other factors are also considered.

The influence of mould conditions on the soundness of castings E. Longden (*Brit. Found.* 1959, **52**, Nov., 477–486) Test casting studies are reported showing the importance of strong, rigid, high-thermal conductivity moulds, especially for grey iron. The use of high P iron for projectile manufacture is described.

The influences of mould conditions on the soundness of castings E. Longden (*Inst. Brit. Found. Australian Branch*, 1959, **10**, 5–23) After a discussion of the factors on which soundness of castings depend, the author describes work on cast-iron shells, with particular reference to the importance of strong, rigid mould walls, and the use of desensers.

An investigation into factors influencing the soundness of ingot-mould shrinkage K. E. L. Nicholas (*Brit. Found.*, 1959, **52**, Nov., 487–494) The shrinkage of ingot-mould castings of 4–140 t weight of low-P haematite iron was studied. The cause was insufficient mould rigidity, also variable metal composition and high pouring temp.

FORGING, STAMPING, DRAWING, AND PRESSING

Simplified anti-vibration hammer foundations W. Eastwood (*Engineer*, 1960, **209**, May 27, 910–911) A letter in reply to criticism of an earlier article.

Development and present position in the finishing of steel shaped parts by cold forging H. D. Feldmann (*Draht*, 1959, **10**, Aug., 397–398) Examples of parts made by cold extrusion and the range of application of the process are given.

Mechanized handling of billets for crankshaft forgings H. Chase (*Machinery*, 1960, **96**, Feb. 3, 242–244) A description is given of plant specially designed for mechanical handling including billet production, stacking transfer heating, and subsequent progressing through forging operations in the forging plant of the Oldsmobile Division, Lansing, Mich., USA.

BISRA demonstrate programmed forging operation (*Process Control Autom.*, 1960, **7**, April, 170) A brief description of completely automatic programmed forging of steel at BISRA's experimental forge at Sheffield is given. Forging was carried out by a precision controlled 200-t press interlocked with a fast remote controlled experimental manipulator capable of longitudinal rotational and lifting movements. Electric controls had been developed for both press and manipulator. The forging schedule was calculated from a theory of the changes in shape worked out in the laboratory.

Nickel process makes patterns without costly machining (*Iron Age*, 1960, **185**, April 7, 89–92) A new process which employs nickel carbonyl for making dies is described. A mould of the die-shape is heated in a depositing chamber and nickel carbonyl gas is passed over it; nickel is allowed to build up on the mould to the required thickness. A mould costs about the same as the machined type but advantages are obtained from speed of production; resistance to abrasion and wear; high thermal shock resistance; smooth, non-binding surface; ease of repair; ease of making intricate shapes. Dies are suitable for forming, drawing, casting patterns, and core boxes.—D.L.C.P.

The sizing of stampings and drop forgings R. Meier (*VDIZ*, 1960, **102**, May 21, 614) An abstract of a paper 'the economic application of cold working in production practice' (*VDI Berichte*, 1959, **39**), dealing with factors influencing surface quality and dimensional stability.

Classification and systematization of sheet metal press-forming process K. Yoshida (*Sci. Papers Inst. phys. Chem. Res.*, 1959, **53**, Sept., 125–187) [In English] Forming processes on sheet, using a metal flat-faced die are discussed theoretically, classified, and methods of testing for 'press formability' are reviewed (49 refs).

The deformation of metals by the use of explosives E. O. Genzsch (*Machinenewelt u. Elektrotechn.*, 1960, **15**, June, 225–228) The forming of metals by the use of explosives and riveting by explosive rivets are briefly explained. Explosives are also used for the surface hardening of steel.—T.G.

Navy expert outlines ABCs of explosive metalworking J. S. Pearson (*Steel*, 1960, **146**, March 21, 140–143) This article describes the fundamentals of shaping and joining metals by explosives, the type of operations that are possible, and the influence of the explosive charges on the materials being worked.

Deep drawing in France G. Chauvelin (*Europäischer Maschinen Markt.*, 1960, **10**, (1), 12e–21e) A review.

Determination of yield in rod and pipe drawing H. Orzechowski (*Rudy i Metale*, 1960, **5**, (3), 105–107) The author gives equations for the theoretical and actual yields. The largest amount of waste in the process of drawing pipes and rods is due to discarded sharpened ends. Consequently modern drawing equipment, giving partly finished products in lengths of about 40 m or equipment using winding drums giving even longer lengths, gives the highest yield.

The drawing of carbon steel wire at high speeds W. Holloway jun. (*Draht.*, 1959, **10**, Aug., 394–396) A description is given of work carried out by Speedwell Wire Co. Ltd, and Barron and Crowther Ltd to establish the best conditions for the manufacture of high C wire at high drawing speeds. The steel was 0·7%C, and the finished wire was from 0·25–1·37 mm dia.

The magnitude and direction of axial forces acting on the workpiece in the cold reducing of tubes Yu. F. Shevakin (*'Stal', *Rolling and Tube Manufacture Supplement*, 1958, 295–313) Equipment designed to measure the magnitude of the axial forces and results obtained are described. Roll pressure, axial force on the work piece, time and travel were recorded on film using an oscilloscope, and forces acting in the roll-workpiece system during rolling are calculated.

Deformations and stresses during drawing R. Levi (*Ing. Mecc.*, 1960, **9**, May, 17–24) In the first part of this paper an attempt at classifying deep-drawing operations is made. In the second part deformations and stresses in the various types of deep drawing are studied including cylindrical, non-cylindrical, and more complex operations.

Quality, production rhythm, and safety in the deep drawing industries J. Dupas (*Mét. Const. Méc.*, 1960, **92**, April, 277–281) A description of the principles, operating diagrams, and specimens of automatic feed distributors, electro-pneumatic, and pneumatic machines ensuring automatic passage of material and parts from one operation to another.—S.H.-S.

Presses in the Federal German Republic for producing tubes of less than 150 mm diameter A. I. Tselikov (*Stal'*, 1960, (5), 449–455) An account of a visit with descriptions of extrusion plant for 25–30 mm dia. tube, welding equipment for 17–19 mm tubes and cold-rolling machines for 1¼–6½ in. steel tubes. Drives are also described and recommendations for Soviet mills are made.

Lone Star Steel Co.'s new stretch reducing mill L. Eller (*Iron Steel Eng.*, 1960, **37**, March, 78–85) Layout and operating details of this plant are given, which permits the fabrication of tubing, and smaller size casing and line pipe from electric welded tube. Extreme flexibility of operation and increased output have been obtained.

The Bohumin Steel and Wire Works –15 years of Socialist working F. Pališek (*Hutnik*, 1960, **10**, (2), 52–55) The development of production and productivity in the wire works and associated enterprises in post-war years is discussed. Productivity in the manufacture of wire and wire rope has approximately trebled, and output has increased about 12 times.

Tension in wire engineering E. J. Saxl (*Wire Wire Prod.*, 1959, **34**, Dec., 1611–1622, 1664–1672) A comprehensive survey of the meaning and practical implications of wire tension is presented. A better understanding is required for a modern appreciation of such subjects as wire drawing and coil winding. Methods of measuring tension and mechanisms to compensate for tension variations, including a new electric tension control, are described.—D.L.C.P.

Design and economic layout of wire twisting plant H. W. Aengeneyndt (*Draht*, 1959, **10**, Aug., 339–345) The production of high-strength twisted wire for concrete reinforcement and the equipment used are described.

Definitions and simplified methods of calculation for wire manufacture A. Mohrnheim (*Draht*, 1959, **10**, Aug., 345–348) Definitions for wire processing are given, and simple mathematical relationships derived. A practical method of solving wire drawing problems using slide rules is illustrated by examples.

Method of recognizing small irregularities in bent wire bodies F. Manna (*Draht*, 1959, **10**, Aug., 348–352) By means of measurements on the elliptical track formed by vibrating one end of a spring steel wire while the other end was gripped, it was possible to identify and fix the principal planes of bending in individual cuttings of the wire.

The manufacture of precision cold drawn shaped wire J. Rigby (*Wire Wire Prod.*, 1960, **35**, Jan., 47–53, 125) Types of shaped wire are described and the method of manufacture outlined. The making of dies and methods of machining carbide are considered.—D.L.C.P.

The modern development of the wire rope D. van de Moortel (*Draht*, 1959, **10**, Aug., 353–357) The construction and operation of a special type of fatigue testing machine for wire rope testing is described, and the results obtained on various rope lays are discussed.

The drawing of high carbon steel wire at high speed W. Holloway jun. (*Wire Wire Prod.*, 1960, **35**, Feb., 199–200, 242) The technique of drawing high-carbon steel wire at speeds up to 3500 ft/min. is described.—D.L.C.P.

Increased strength of steel ropes V. P. Emel'yanov (*Metallurg.*, 1959, (12), 32–33) Suggestions made at 'Zaporoshstal'' to improve the product are listed, the chief of which is that care should be taken to give a uniform tension to the strands and twists.

Steel ropes—an example of effective scientific and technical co-operation Z. Steininger (*Prz. Techn.*, 1960, **81,** (15), 17) The author gives the title of papers presented to a conference of representatives from Hungary, East Germany, Rumania, USSR, and Czechoslovakia dealing with the possibilities of increasing the life of steel ropes. Each country specializes in certain directions and a free exchange of results enables full use to be made of the experience of other countries.

Extrusion of metals S. de Filippi (*Ing. Mecc.*, 1960, **9,** Feb., 63–68) The article outlines the application of the mathematical theory of plasticity as the Sachs formula does not take the internal deformation work into account. An experimental method is described by which all the deformations can be evaluated and the required quantities are obtained from the deformations by applying the theory of plasticity.

The extrusion of steel to tubes and sections E. K. L. Haffner and R. M. L. Elkan (*Neue Hütte*, 1960, **5,** May, 266–278) The range of application of extrusion to steel is compared with rolling processes, and its advantages in the manufacture of tubes and profiles in austenitic, ferritic, and high-alloy steels are shown; a comparison of costs is given. Some details of plant design are given, and problems connected with the construction of modern presses and the use of glass lubricant are discussed.

ROLLING MILL PRACTICE

Steel research center revamps rolling mills (*Steel*, 1960, **146,** March 14, 102) Refinements to the equipment in the rolling mill laboratories of US Steel Corp. are described.

Determining the contact surface by calculating the elastic compression of the rolls A. A. Dinnik (*Izvest. VUZ Chern. Met.*, 1959, (10), 71–78) The important effect of the size of the arc of contact and of the angle of bite on the magnitude of elastic compression of the rolls has been experimentally demonstrated and a method of determining the increment of the angle of bite because of the elastic compression, under the actual conditions of rolling of the steel and chilled cast iron rolls in given, either by calculation or graphically.

Investigating the pressure conditions in the continuous small section mill V. N. Vydrin, M. E. Boiko, P. N. Amosov, and S. I. Moshkin (*Izvest. VUZ Chern. Met.*, 1959, (6), 37–42) A new method is proposed for calculating the pressures for the continuous rolling, which does not require loop shaped strips and is therefore of universal application. The necessary equipment has been evolved and tested under conditions of production. The relation between pressure and rate of rolling is of a linear character.

Automatic measurement of the rolling force as a basis for an efficient schedule of reduction M. A. Zaikov, V. S. Tseluikov, V. M. Permyakov, and G. G. Tereshin (*Izvest. VUZ Chern. Met.*, 1959, (6), 53–62) Work extending over a considerable period at the Kuznetsk Metallurgical Combine has resulted in the design of a reliable instrument for automatically recording the rolling force. A description of the tensometer is given, also examples for determining the rolling conditions for different types of steels.

Simulation of the dynamics of sheet rolling on a continuous mill S. N. Kozhevnikov and A. P. Pukh (*Izvest. VUZ Chern. Met.*, 1959, (6), 133–144) Theoretical consideration of the work of two adjoining stands of a continuous rolling mill compared with experimental data obtained on an electronic model at the TsNIITMASh have shown that the calculated data are sufficiently near to the dynamics of the process. The method of reproducing the actual process may be used for the laboratory investigation of various control systems for the continuous cold-rolling of sheet.

Geometrical considerations in oblique rolling P. K. Teterin (**Stal'*, *Rolling and Tube Manufacture Supplement*, 1959, 243–258) Solutions are derived which are applicable to conical, barrel-shaped, or disc rolls.

The hardening curve in the hot rolling of steel M. A. Zaikov, Ya. V. Shamets, and V. N. Peretyat'ko (*Izvest. VUZ Chern. Met.*, 1959, (9), 73–82) Present-day calculation of rolling forces take into account only temp. and the scheme of the rolling forces, rarely the rate of deformation and the chemical composition of the metal. The deformation ratio is introduced into the equation as a component of the system of forces, but it does not determine the hardening during the deformation. Experiments are described on a laboratory rolling mill fitted with a bear box allowing a change of rolling speed from 0·2 to 6·1 m/sec. and equations are evolved for plotting a hardening curve in the hot rolling operation. The metal temp. has only a weak influence on the factors investigated and may be neglected at a first approximation.

Hot rolling of steel C 35 with heavy reductions Z. Wusatowski, W. Gorecki, and E. Judycki (*Neue Hütte*, 1960, **5,** May, 257–266) Following early trials with mild steel (*Berg–Hütten, Monatsh*, 1958, **103,** 136–142, 153–160) three sets of trials were carried out on C 35 steel, with reductions of from 50–90%. The results are reported and analysed.

Roll design for reducing and sizing mills A. P. Chekmarev and G. I. Gulyaev (**Stal'*, *Rolling and Tube Manufacture Supplement*, 1958, 276–294) In rolling small-dia. tubes without the use of a mandrel, the authors have found that an oval is the most suitable shape for the passes of sizing and reducing mills. Calculations for the pass design are developed, and applied to the rolling of plain carbon, low-alloy, and stainless steel tubes.

The production of iron rolls for rolling mills. I D. Magri (*Fond. Ital.*, 1960, **9,** (2), 49–58) This first article summarizes the processes by which iron mill rolls are manufactured. Their general characteristics are defined and the influence of the various chemical elements is described. The aim is to give an account of the considerable problems involved in their manufacture. **Part II** (3), (95–108) A detailed description is given of the characteristics and manufacture of various types of semi-hard rolls. There are tables classifying the various types of rolls, giving chemical analyses and the resistance values of test pieces.

Rolls for rolling mills. Part I R. A. Villanueva (*Ing. e Indust.*, 1958, Nov.–Dec., 78–84) Materials used in the manufacture of the rolls are enumerated, and a general description of the composition and structure of grey cast iron rolls, with and without alloy additions, is given. **Part 3** (1959, March, 87–89) In this part the constituent elements in Fe, divided into 'common' (C, Si, Mn, S, P) and 'alloy' (Ni, Cr, Mo, V, Cu, Zn, Al, Ti, Co, Te, B) are discussed in relation to their effect on iron properties. **Part 4** (April, 88–94) A brief reference is made to oxidation losses of various elements in melting, and this is followed by a discussion of the effects of composition on chilling, and of the method of testing for depth of chill. Then a description is given of the mould for the manufacture of chill rolls, and various defects in manufacture are enumerated and discussed. **Part 5** (Sept., 131–134) The discussion on manufacturing defects is continued, and is followed by a description of typical compositions of unalloyed, Mo and Ni cast iron rolls, and their heat treatment. **Part 6** (1960, **27,** April, 90–93) The effects of alloy additions on chill roll structure are outlined, and the applications of these rolls, classified as plain C, Mo-containing, and Ni-containing, are reviewed.

Increasing the life of steel rolls of Pilger type tube mills O. Pejčoch (*Hutník*, 1960, **10,** (2), 65–68) Experience with repairs to rolls by automatic welding, obtained in the Klement Gottwald Steelworks, is discussed. The method is shown to facilitate reducing expenditure on rolls by 48%.—P.F.

Grooving the rolls of pilger mills to ensure maximum productivity O. A. Plyatskovskii and A. A. Chernyavskii (*Izvest. VUZ Chern. Met.*, 1959, (10), 49–56) A method is suggested for calculating one of the principal parameters by taking into consideration the characteristics of the feeding unit. The resulting calibration ensures increased productivity and an improvement in the quality of the rolled tubes.

Carbide ups roller guide life International Nickel Co. (*Iron Age*, 1960, **185,** Feb. 11, 158) The advantages of using carbides for components of roller guides in a 9 in. wire mill finishing stand are described. The rollers are made of 'Kentanium' TiC alloy and the bearing pins from 'Kennametal' WC alloy. Advantages over lubricated steel bearings include: life increased 20 times; maintenance costs reduced 80%; mill capacity increased; product quality improved; lubrication not required and seizing avoided.—D.L.C.P.

Fire cracking in rolling mill rolls N. R. Arant (*Iron Steel Eng.*, 1959, **36,** Nov., 100–107) Theories of fire cracking are described and assessed with reference to the literature (18 refs).

Determination of an economic stock of mill work rolls G. D. Feigin (*Stal'*, 1960, (4), 342–345) An operational research study of the optimum stock of spares for a rolling mill to ensure uninterrupted operation and also minimum consumption is presented.

Maintenance of mill oil bearings C. A. Bailey and A. C. Cooper (*Iron Steel Eng.*, 1959, **36,** Nov., 65–84) Design features, maintenance, and lubrication of these bearings are discussed in great detail, the article being illustrated by a number of diagrams and photographs. Two-high and four-high mills are dealt with.

Spray lubrication of roll necks A. A. Paul and G. C. Almasi (*Iron Steel Eng.*, 1960, **37,** Feb., 102–109) Results of experience in this method of lubrication are described, and its advantages assessed. Longer bearing life is achieved, shop work costs are reduced, refinishing, and dressing operations are diminished, mill lining delays reduced to a minimum, and consumption of lubricant is less.

Report of the enquiry on blooming and slabbing mills, May–Sept. 1957. Part 2—layout of passes, rolling procedure, roll knurling and spraying C. Roger (*Centre Doc. Sidér. Circ.*, 1959, (4), 927–950) Practices at a number of works are tabulated in detail, and discussed and compared critically.

Investigations of failures in a blooming mill F. E. Stacey (*JISI*, 1960, **196,** Oct., 195–207) [This issue].

From ingot to ingot in fifteen days—replacing a 27 in. universal slabbing mill with a 38×66-in. slabbing mill on the same site D. H. Samson (*Iron Steel Eng.*, 1959, **36,** Oct., 102–112) The planning and preliminary work of replacing an old two-high mill by a modern two-high reversing slabbing mill at the Dominion Foundries and Steel Ltd., Hamilton, Ontario, were so complete that only a 15-day shutdown was required. Details of the work are given.

Ingenious construction minimizes production delays H. Ellsberg (*Iron Steel Eng.*, 1959, **36,** Oct., 154, 157) An account of the structural alterations carried out in such a way as to keep a rolling mill in operation while replacing a two-stand billet mill with a 30-in. six stand mill.

Problems of the complex automation of a blooming mill Z. Wasowicz (*Hutnik*, 1960, **27,** (2), 47–55) The author gives the results of measuring those parameters in a mill which by the introduction of electro-mechanical control of particular parts of the mill will reduce idle time in the rolling process. This saving of time is accomplished by increasing rolling speed, decreasing the period of time between one operation and the next, eliminating the human operators' reaction time and reducing the time necessary to reset the rollers. The relationship between the time necessary to reverse the ingot and the time necessary to reset the rollers makes automatization almost imperative. At the very least the main drive, conveyor drives, and roller resetting drives should be automatically controlled.

Automatic blooming mill screwdown control systems C. Roger (*Centre Doc. Sidér. Circ.*, 1959, **16,** (4), 903–925) A detailed description of various systems applied in Great Britain, USA, and USSR.

Rational finishing passes of a 300 mm mill for rounds of 14–32 mm diameter N. I. Tereshenkov (*Stal'*, 1960, (3), 253–254) Redesign eliminating the bevel edge imprint previously encountered is outlined.

Two-stand rolling of concrete reinforcement bar with the use of repeaters A. F. Minaef, A. A. Nefedov, N. V. Telushkin, N. S. Ter-

minosyan, A. I. Kurilov, L. N. Skachkov, M. M. Polyakov, and I. A. Lipovetskii (*Stal'*, 1960, (3), 234–243) An account of the reconstruction of the 280 mm mill at the Enakievo Works and of the development and planning of the equipment and controls. Considerable savings are claimed and savings of labour.

The rolling of round wire into flats and rectangles A. F. Sperduti (*Wire Wire Prod.*, 1959, **34,** Oct., 1300–1305, 1409–1411) Factors governing the selection of the mill, the electrical drive, and mill accessory are discussed with illustrations of examples.

The fully continuous rod mill and its operation T. Ogura and Y. Yoshiyagawa (*Tetsu to Hagane*, 1959, **45,** Sept., 978–979).—K.E.J.

Rheinhausen wire rod mill—a modern continuous single-purpose plant W. Schulz (*Wire*, 1959, (42), Aug., 94–101) Layout and operation of the billet reheating furnace, roughing, intermediate, and finishing trains, coiling and auxiliary equipment, and drives and control mechanism are described.

Controls on flat-wire mill insure fast versatile output H. C. Swartfiguer and R. K. Larson (*Iron Age*, 1960, **185,** Jan. 21, 88–89) A general account is given of the method of control of a new unit at GEC.

Economic rolled profiles J. Smolik (*Hutn. Listy*, 1959, **14,** (12), 1049–1060) Some experiences of Soviet production and projects of development of beams and sections of new designs and lighter weight are discussed with particular reference to possibilities of improved designs of weight-saving girders and sections for Czechoslovak industry. The design of comparable sections in various countries is considered and their respective 'economic indices' are evaluated.—P.F.

Quality control in an alloy steel bar mill T. P. Davis (*Wire Wire Prod.*, 1959, **34,** Oct., 1289, 1292, 1393–1394) The quality control system at the bar mill of Allegheny Ludlum Steel Corp., Watervliet, NY, is described.

Contour rolling (*Met. Prog.*, 1959, **76,** Oct., 125–127) A general outline of the scope and advantages of the process, some applications being enumerated.

A more accurate method of calculating roll pass designs for angle irons B. P. Bakhtinov (*Stal'*, 1960, (4), 330–333) Applications to a new Russian specification for rolled products are described.

A research on the change of mechanical properties in the rolling of heavy rails Chang Tso-mei, Gau Tze-ming, and Lei Ming-foo (*Acta Met. Sin.*, 1959, Sept., **4,** 222–234) In trapezoidal passes, changes in the height and angle of the cutting-in wedge have little effect on the properties of the foot of the rail; hence wedges with small height and large angle are best. The use of two or four such passes can be chosen on production grounds. To give good properties, when the cooling rate is high, the rolling temp., reduction, and rolling speed can be increased to reduce the number of passes used (11 refs).—K.E.J.

The modern technology of rolling heavy rails R. Miculescu (*Met. Constr. Masini*, 1959, (11), 934–939) A general description of the rolling process of heavy rails which consists usually in 9–11 passes of which 2–4 are preliminary, 2–4 intermediate, and 3–5 final passes. The arrangement of grooving is given.

Modern technology of rolling heavy rails R. Miculescu (*Met. Constr. Masini*, 1959, (12), 1033–1037) Suggestions are made regarding cooling after rolling, heat treatment, also regarding the design and construction of new rolling mills.

The economics of a continuous light-section rolling mill W. A. Antoni and E. Schneider (*Stahl Eisen*, 1960, **80,** May 12, 641–652) The economic principles of operation of a continuous light-section rolling mill are considered. Based on such considerations a continuous rolling mill for light sections was built at Hoesch–Westfalenhütte, Dortmund, and its operation and the difficulties involved in its economic efficiency under prevailing German conditions are discussed. The individual quantities ordered by the customers are still too small for really 'continuous', i.e. round-the-clock, operation, and downtimes are still too long. It is suggested that higher rebates for larger orders than have hitherto been customary should be offered.—T.G.

A study of the sheet mill fabrication of metal A. V. Emel'yanov and A. M. Meleshko (*Stal'*, 1960, (4), 350–353) Calculations on discard and rolling losses are given, and these are related to sheet size and thickness. Minimum coefficients are arrived at making possible more economical operation.

Grooving universal finishing ovals for rolling discs M. L. Mirenskii (*Izvest. VUZ Chern. Met.*, 1959, (10), 63–70) The dimensions of the finishing oval groove must be calculated on the basis of the maximum productive power of the given mill. The shape of the oval is determined geometrically. The box-like shape ensuring a more intensive deformation and a cleaner surface of the oval strip is preferred. The universality of the finished oval groove depends on the minimum diameter of the disc. The maximum dia. is determined by the admissible minimum deformation in height of the oval strip in the round groove.

The accuracy of the rolled material and the rigidity of the stands A. P. Chekmarev and G. G. Pobegailo (*Izvest. VUZ Chern. Met.*, 1959, (8), 69–74) The improvement in the accuracy of rolling goes together with the improvement in the technology of rolling and with the improvement in the construction of the rolling mill stands. The stands for the finishing mill must have a minimum elastic deformation, in any case not exceeding the order of tolerance for the rolled material.

Backup rolls and their contribution to gage variation E. G. Lindsey (*Iron Steel Eng.*, 1960, **37,** March, 91–98) Improvement obtained by applying scientific control to roll shape and quality, particularly by improving roll concentricity, is discussed.

An efficient arrangement for a blooming mill L. V. Andreyuk (*Metallurg*, 1959, (12), 24–28) At the Magnitogorsk Combine work has been carried out at two blooming mills during the last five years on improving the layout and equipment. Production has meanwhile increased by 20% at mill No.2 and by 42% at mill No.3. The scheme for the installation of the equipment is briefly explained with diagrams.

Efficient method of rolling ingots on the breakdown stand M. I. Kapustina, I. D. Kuzema, B. S. Kirillov, V. D. Danilov, and A. M. Savchenko (*Izvest. VUZ Chern. Met.*, 1959, (9), 95–100) An efficient method is described of roughing on the one-stand two-roll reversing 830 mill working 5/600 × 160/220 mm slabs to 400 × 150 mm and to 140 to 205 mm square blanks with steam driven rollers of 50 KhN steel.

The need for continuous improvement in the quality of steels and rolled sections E. Traian (*Met. si Constr. Masini*, 1959, (11), 374–376)A general consideration of the shortcomings and some suggestions as applicable to the conditions of production.

Manufacturing costs of the rolling of high-grade steel rod as a result of throughput and quantity G. Ibach (*Stahl Eisen*, 1960, **80,** May 26, 744–751) The internal accountancy research was conducted on a new and an old rod mill in which a number of grades of steels were rolled. It was found that the dimensions and the type of steel are the two main factors in the manufacturing costs, the latter being dominant. For some of the steels, dimensions and quantities were determined that are required for an economic operation of the mill.

Roller entry guides for rolling rounds N. F. Gritsuk and V. P. Polushkin (*Stal'*, 1960, (4), 338–342) The designs and advantages of the guides are set out. Materials for their construction and wear life are indicated.

Grooving for an equal sided steel angle with large reduction D. I. Starchenko (*Izvest. VUZ Chern. Met.*, 1959, (9), 83–94) Calculations are given for roughing, intermediate, and final grooving of these sections, and examples of rolling angle 40 × 40 × 4·5 mm in three passes from 30 × 30 mm blanks using the diagonal method with reduction ratios of 1·695, 1·936, and 1·278 are added.

Electronic control of rolling mill operation M. Cliques (*Tech. Mod.*, 1960, **52,** April, 197–199) A description is given of the operation of electronic equipment and its application to the selection of sheets on the shearing line.

Giant plate rolls for G.E.C. (*Weld. Met. Fabr.*, 1960, **28,** March, 103–105) The new three-roll cold plate bending machine recently installed at the Fraser and Chalmers Engineering Works Erith, is illustrated and briefly described.

A modern high speed tandem cold reduction tinplate mill N. W. Hocking jun. and A. F. Kenyon (*Iron Steel Eng.*, 9160, **37,** March, 111–120) The main features of this type of mill are described, and the problems connected in fitting such a mill into existing buildings are discussed. The need for automatic gauge control is emphasized, and some results obtained with gauge control are described and illustrated by strip thickness charts.

A fresh approach to automatic gage control on rolling mills (*Iron Steel Eng.*, 1960, **37,** March, 125–132) Work carried out by the electromechanisms department of Allegheny Ludlum Steel Corp. is described in which reduction measurements based on strip elongation measurements and automatic logic control with a computer are used as a basis for gauge control.

Gauge thin strip accurately J. A. Buza (*Iron Age*, 1960, **185,** Feb. 18, 120–121) A system of measuring the thickness of steel foil to millionths of an inch is described. The thickness of the sample is compared with the wavelength of a Kr light source.—D.L.C.P.

Modern data acquisition system classifies tinplate coils Jones and Laughlin Steel Corp. (*Iron Age*, 1960, **185,** March 17, 112–113) The use of an Automatic Inspection Data Accumulator (AIDA) for recording the quality of tinplate coils as they are produced, is described. Information from instruments and visual inspection is fed continuously to the AIDA and recorded on punched cards. Data on the following is recorded: strip thickness, pinholes, coating thickness, length, weight. A record of the quality of large coils is thus available.—D.L.C.P.

A comparison of pneumatic and hydraulic braking devices for Pilger mill feed mechanisms V. S. Artyukh (**Stal'*, *Rolling and Tube Manufacture Supplement*, 1958, 338–351) A mathematical analysis of the two types of mechanism shows that hydraulic braking has several advantages over air braking, including greater efficiency.

Metal pressure on components of an electric resistance-weld tube mill Yu. A. Mednikov and G. Ya. Gun (*Stal'*, 1960, (5), 445–449) Pressures between metal and rolls and metal and electrodes in the weld pass were measured and calculations were made to derive optimum conditions and the best designs for weld components.

The influence of roll calibration in draw rolling on the mode of plastic deformation and on the reject rate in tube production O. Pejčoch (*Hutn. Listy*, 1960, **15,** (2), 157–166) Longitudinal faults and cracks on the tube surfaces were found to originate at inclusions, and to be induced by high shear stresses during rolling. These shear stresses can be reduced by modifying the cone angles of the rolls and piercing cone.—P.F.

Faults occurring in the production of seamless tubes O. Pejčoch (*Hutník*, 1960, **10,** (1), 13–20) A detailed survey is made of the nature of defects, and their origins and prevention are discussed. Illustrations are taken from the author's researches on the problem, carried out in the Ostrava Steelworks. The main cause of defects is heterogeneities in the steel.—P.F.

The chemical composition of steels used in making piercing tools in tube rolling and the possibilities of heat-treating it R. Šejnoha (*Hutník*, 1960, **10,** (1), 5–9) The mandrels are of ledeburitic steel containing 2%C, 20%Cr, 1·5%Ni, 0·5%W, and 0·2%V. They are used without heat treatment. It is shown that the life of mandrels can be substantially increased by tempering for at least 2 h at 800°C.

Sheet and plate making machinery (*Mét. Contr. Méc.*, 1960, **92,** April, 221–229) The range of sheet and platemaking machinery comprises cutting, bending, straightening, and folding machines. The products of the principal manufacturer in each group are described and illustrated by examples of modern equipment.

MACHINERY AND SERVICES FOR IRON AND STEEL PLANT

The water economy of an integrated iron and steel works E. Knecht (*Stahl Eisen*, 1960, **80,** May 26, 751–756) The conversion of the water economy of an integrated iron and steel works to re-use of the water is described and the advantages—including substantial savings effected by the drastic reduction in purchased water—are pointed out.—T.G.

System stacks steel quickly North-Western Steel and Wire Co. (*Iron Age*, 1960, **185,** Jan. 28, 99) An account of an automatic handling system for movement from rolling mill to loading bays.

Benson boiler at steelworks (*Elect. Times*, 1960, **137,** June 16, 961–963) The installation fired with blast-furnace gas is described and illustrated.

Automatic control at Margam marshalling yard (*Engineer*, 1960, **209,** April 22, 679–680) A general illustrated account.

Electric overhead bridge cranes for steelworks S. Bellone (*Fond. Ital.*, 1960, **9,** (2), 59–64) The maintenance of means of lifting and transporting, particularly electric bridge cranes, is a very serious problem in steelworks and steel foundries. To reduce maintenance costs and above all to reduce breakdowns and stoppages, special planning methods must be adopted for these installations. This article describes the criteria used and the characteristics necessary for these cranes.

Steelworks cranes from the work study point of view K.-H. Prolingheuer (**Fördern und Heben*, 1958, (8), Feb., 58–61) A discussion of crane cab design with the object of alleviating the work of the driver. Points discussed are: design to give an uninterrupted and comprehensive view of the working area, to give a comfortable working position, and to enable the controls to be operated easily and rationally.

EOT crane runway rail installations W. J. Tunny (*Iron Steel Eng.*, 1960, **37,** March, 67–72) Factors involved in poor crane rail performance are examined, and remedies proposed. These include rail and crane wheel design, and methods of installing and joining rails.

Crane bridge drive and wheel combinations R. F. Dehn (*Iron Steel Eng.*, 1960, **37,** March, 73–75) A review of various combinations in general use, with a discussion of some problems involved.

The selection of crane wheels R. B. McCready (*Iron Steel Eng.*, 1960, **37,** March, 76–77) A brief review of the various types of crane wheel materials and design, and a discussion of the application of the Hertz formula for calculating allowable wheel loads.

The cranes of the iron and steel works Rourkela in India H. Rieländer (*Stahl Eisen*, 1960, **80,** May 26, 756–763) A description of the various types of cranes that are used in this new iron and steelworks is given. There are all together 73 cranes, ranging from 5 to 150 t capacity. Transport of the dismantled cranes to India and their erection there is included in this description.—T.G.

Gas turbine plants with free-piston gas producers and after burner for economic use during peak hours in integrated iron and steel works J. W. Jacobi (*Stahl Eisen*, 1960, **80,** June 23, 894–896) The principle of this type of gas turbine is briefly explained and its usefulness for peak-hour boosting of power pointed out.

Establishing a table of tariff-rates in agreement with the actual cost of transport in a works' own railway system W. Marquardt (*Stahl Eisen*, 1960, **80,** June 9, 809–818) A somewhat simpler system for arriving at the true cost of internal railway transport in iron- and steelworks is derived.—T.G.

LUBRICATION

The lack of a lubrication engineer in the drawing office as the main cause for breakdowns of journal bearings K. Droste (*Stahl Eisen*, 1960, **80,** May 26, 766–768) The fundamentals of journal bearing lubrication are set out, illustrated by a number of diagrams of the oil-pressure distribution in these bearings. The causes for breakdowns during service are discussed.—T.G.

Molybdenum bisulphide, an aid for the rationalization of machining Handwerkstechnischen Institut (*Maschinenwelt u. Elektrotech.*, 1960, **15,** June, 229) The merits of MoS_2 additions to cutting oils in the machining of steels are pointed out. An addition of 5%MoS_2 brought about an increase of life before regrinding of 100%.

WELDING AND FLAME-CUTTING

Spotlight on welding (*Met. Prog.*, 1959, **76,** Dec., 123–124) A note on points raised at the Autumn, 1959, meeting of the American Welding Society, on reproducing spontaneous brittle fracture in laboratory specimens, and on recently developed welding and inspection methods.

Welding to-day E. C. Pigott (*Iron Steel*, 1960, **33,** May, 185–187) Classification, shipyard welding, use of radiography for testing, and features governing design are reviewed briefly.

Welding and cutting techniques in the manufacture of metallurgical products W. Gilde (*Neue Hütte*, 1960, **5,** April, 232–237) Subjects dealt with include electroslag welding, repair welding, building up small parts by electroslag welding, and powder and argon flame cutting.

Link automation to CO_2 Welding Borg and Beck (Borg Warner Corp.) (*Iron Age*, 1960, **185,** Feb. 18, 117) Arcweld equipment which has been converted so as automatically to spot-weld a gear to a clutch housing is described.

Welding nodular cast iron R. C. Bates (*Met. Prog.*, 1959, **76,** Nov., 95–99) It is shown that a suitable annealing treatment improves ductility in arc welds by elimination of carbides. Investigation of transition temp. on the Linde Co. technique using oxyacetylene welding showed that no carbide was formed except in the root area, but that transition temp. were high.

The preparation of welds for shielded submerged arc welding M. Mosný (*Zváranie*, 1960, **9,** (2), 42–45) Preparatory work required to assure sound welds is discussed.—P.F.

The technology of welding high strength chain links M. Vltavský (*Zváranie*, 1960, **9,** 3, 74–80) A detailed analysis is made of the technology and economics of chain link welding, welding machines, and presses used in Czechoslovakia for this purpose are considered, and tests designed to ascertain optimum welding conditions are discussed.—P.F.

Three step welding method speeds repairs on castings (*Iron Age*, 1960, **185,** April 7, 96–97) A new three-step method of repairing castings is described; it is fast and avoids cracking and distortion. In the first step a chamfering electrode is used to make a V which is then grooved to afford stress relief. The area is then covered with material from a sealing electrode which protects and prepares the surface. Finally the area is welded up using a back step technique; quenching is effected simply between each pass.—D.L.C.P.

Welded design offers savings in machine construction (*Iron Age*, 1960, **185,** March 24, 92–94) Examples of how welded construction has replaced cast components for machine tool bases are given. Providing the unit is completely redesigned considerable cost savings are possible.—D.L.C.P.

The production of tubes of large diameters from steel 19G A. M. Garagulya (*Stal'*, 1960, (4), 346–350) An account of th automatic submerged arc-welding process at the Khartsyst works, the improvements introduced, and the problems still to be met. The 19G steel is relatively cheap and has been substituted for the 14KhGS steel previously used.

PROPERTIES AND TESTS

Measuring the relaxation period in the polymorphic transformation of iron R. I. Garber and A. I. Kovalev (*Fiz. Met. Metalloved.*, 1959, **8,** (5), 785–788) The damping decrement of torsional oscillations of cylindrical specimens of technical iron with 0·04%C has been measured between 20 and 1000°. In the region of the $\alpha \rightarrow \gamma$ transformation this decrement was measured for five frequencies from 1·08 to 6·8 cycles. The frequency relation of damping at the transformation temp. shows a maximum with 2·6 cycles which corresponds to a relaxation period of 0·06 sec. Assuming that damping is related to an activation process and that the approximate frequency of the oscillations of atoms is 10^{13} sec.$^{-1}$ the value of the activation energy may be calculated. It corresponds to 64 kcal/g. atom.

The electrolytic production of sulphur-free iron W. T. L. Neal, J. F. Duke, and H. R. Sullivan (*JISI*, 1960, **196,** Oct., 131–134) [This issue].

Onset of relaxation in the spin system in some compounds of iron group elements N. S. Garif'yanov (*Zhur. Ekspt. Teoret. Fiz.*, 1958, **35,** (2), 530–532).

Contribution to the complete determination of the state of residual stress in hollow cylinders P. Lauret, Tran-Xuan-Thoai, and G. Altmeyer (**Forschung auf den Gebiete des Ingenieurwesens*, 1959, **25B,** (2), 44–54) A mathematical proof of the Sachs equations is given and tests on brass and copper and also on steel showed annealing and quenching to have no effect on residual stress level. The method of strain gauge measurement used is said to obviate the need for boring and turning tests and give results in good agreement.

Dislocations in α-iron D. G. Brandon and J. Nutting (*JISI*, 1960, **196,** Oct., 160–166) [This issue].

The geometry of surface cracks P. G. Morgan (*Appl. Sci. Res.*, 1960, **9A,** (2–3), 148–152) The stresses in surface coatings such as enamels are considered.

Thermodynamical calculation of phase diagrams J. L. Meijering (*Philips Separaat*, 2790, *reprint from Phys. Chem. of Metallic Solutions and Intermetallic compounds*, 1958, **2,** June, 1–20; *Nat. Phys. Lab.*, paper 5A) (47 refs).

On the equilibrium between [Mn] and [O] in molten iron K. Gunji and S. Matoba (*Tetsu to Hagane*, 1959, **45,** Sept., 926–928).

Solubility and chemical activity of oxygen in metallic alloys V. V. Averin, A. Yu. Polyakov, and A. M. Samarin (*Izvest. AN Metallurg. i Topl.*, 1959, (1), 13–21) The solubility and chemical activity of oxygen in metallic systems does not change additively through the entire range of concentration of components but depends on the composition of the oxide phase in equilibrium with the alloy of any given composition. Experimental data made approximate calculations possible of the solubility and chemical activity of a saturated oxygen solution in a pure component at a given temp. The nature of the change in the solubility of oxygen in alloys depends on a number of factors, the most important of which is the ratio between reversible dissociations of the oxides of the various components.

The solubility of oxygen in liquid iron containing sulphur A. M. Samarin and Z. Bóžek (*Hutn. Listy*, 1959, **14,** (11), 926–928) An experimental study of the solubility of O_2 in iron containing 0·0–5%S, in the range 1550–1600°C, showed that, contrary to the results of Hilty and Craft (*J. Met.*, 1952, 1307–1312) S has no effect upon the solubility of O within the limits of experimental error.—P.F.

The activity of phosphorus in liquid iron M. G. Frohberg (*Stahl Eisen*, 1960, **80,** April 28, 623–624) The author summarizes recent work on the activity of P in liquid iron. The results are shown in a number of graphs and tables.

Transfer of carbon into solid solution during induction heating of chromium steels I. N. Kidin (*Proizvodstvo i Obrabotka Stali i Splavov; Moscow Steel Inst. Coll. Papers No.*38, 1958, 405–419) Steels with 0·6, 0·8, and 1%C and 2·5, 5·0, and 10%Cr were examined with X-rays to establish heat-treating practices. The C content of martensite and quenching temp. for various heating rates was observed and recommendations for heat treatment are made.

Effect of molybdenum, tungsten and copper on the solubility of graphite in molten iron T. Mori, K. Akita, H. Ono, and H. Sugita (*Tetsu to Hagane*, 1959, **45,** Sept., 929–930) Data are given for the solubility at 1550°C of graphite in Fe–C–W, Fe–C–Mo, Fe–C–Cu,

Fe–C–Mo–W, Fe–C–Mo–Cu, Fe–C–W–Cu, and Fe–C–Mo–W–Cu alloys.—K.E.J.

Simple laboratory bulge tests for sheet-metal fracture-transition studies and for weldment-performance evaluation P. P. Puzak and J. Stoop (*NRL Rept.* 5347, 1959, Aug., pp.15) Rubber is used to transmit the blow to 6×6 in. specimens. Results on steels are given. The method is suitable for routine use.

The mechanical properties of some tempered alloy martensites J. M. Capus and G. Mayer (*JISI*, 1960, **196**, Oct., 149–158) [This issue].

Standard round tensile test pieces (*Annual Rept. Defence Standards Laboratory, Commonwealth of Australia for the year ended June 1959*, 56–57) Work is proceeding on the revision of acceptable tolerances in the machining of test pieces to B.S. 'Tensile Testing of Metals' specifications, and also to determine the effect of lack of coaxiality of the end fittings and the test cylinder on the measured values of the tensile properties of various metals, and the largest size test-piece that can be made from bar steel stock.

Fracture testing of high-strength sheet materials: Testing temperature and effect of temperature on fracture toughness (*ASTM Bull.*, 1960, Feb., 18–28) Research on the standardization of an edge-notch shear test is described in detail. It is concluded that characteristic fracture toughness is the best available test for brittle fracture resistance in sheet material. Edge notch and centre-notch tests are recommended. An instrumental bend test is also considered.

Machining of notched tension test specimens J. L. March, W. J. Ruprecht, and G. Reed (*ASTM Bull.*, 1960, Feb., 52–55) A process of machining in quantity reducing processing time to 1 man-hour each is described.

A consideration of the structure and tensile strength of vacuum-melted Fe C Si system cast iron M. Kikuchi (*Imono*, 1959, **31,** Aug., 712–723) Compositions to give graphitization with no free cementite are found. Vacuum-melted iron had a tensile strength of 30–40 kg/mm², in the carbon steel range a graphite and pearlite structure had 50–85 kg/mm².

Discussion of paper on effect of specimen taper on determination of elongation in the tension test F. H. Dill and E. B. Kula (*ASTM Bull.*, 1959, (241), Oct., 37) A note on standard specimen shape.

Tension test for thin sheet steels G. E. Selby (*Met. Prog.*, 1960, **77,** March, 85–86) Techniques and precautions to be taken in tensile tests on steel sheet up to 0·01 in. thick are described.

Effect of the purity of iron on the discontinuity in the elastic limit observed in tensile curves B. Migaud and J. Talbot (*Compt. Rend.*, 1959, **249,** Nov. 16, 2071–2072) Zone melted iron, like electrolytic iron, shows the discontinuity only if tempered at a temp. at which the C is dissolved before cold-working and ageing.

Contribution to the study of the yield point in wrought iron aged after deformation R. Pankowski (*ATB Mét.*, 1959, **1,** (8), 207–216) Theories explaining the existence of the yield point in the stress–strain curve of recrystallized mild steels and wrought irons are discussed, and an explanation is given in the light of new tests on wrought irons. It is suggested that two separate stages may occur in the ageing of wrought iron, firstly a rapid process related to the anchoring of dislocations and observable in the tensile test, and subsequently a slower process to be seen only in the impact test (11 refs).

The process of formation of 'bright spots' in axle steel N. V. Keis, I. Ya. Aizenshtok, A. I. Komissarov, G. E. Mysina, and E. L. Korotkovich (*Stal'*, 1960, (4), 363–365) Bright spots are observed on tensile fractures. They are shown to be surface tears occurring at points of low ductility caused by heterogeneity in segregation zones.

Examination of material for oil casing pipes with a minimum yield point of 98 kg/mm² R. Šejnoha (*Hutn. Listy*, 1959, **14,** (9), 829–840) The influence on mechanical properties of C, Mn, Si, and Cr contents in quenched and drawn bars of steel with 0·18–0·32%C, 0·97–1·4%Mn, 0·40–1·40%Si, and 0–1·90%Cr was examined statistically. Recommended compositions and treatments are given (21 refs).

Some observations on the behaviour of non-metallic inclusions in steel during tensile testing T. Okada (*Tetsu to Hagane*, 1959, **45,** Sept., 1050–1051) For various steels with given heat-treatments, microstructure, and types of inclusion are correlated with hardness and stress-strain values. Ratios of inclusions separated at the inclusion-matrix boundaries and cracked in the inclusions are given.—K.E.J.

Influences of cementite particle dispersion on the strength of tempered carbon steel G. Mima (*Tetsu to Hagane*, 1959, **45,** Sept., 1055–1057)

Approximate conversion values for the elongation at fracture of steel (*Aciers Fins. Spéc.*, 1960, March, 78) Approximate comparative values established as a result of statistical research are tabulated for different elongations at fracture with different gauge lengths. Rules for using the table are given and its limitations are indicated.—R.P.

Behaviour of steels at low temperatures under alternating tensile stresses M. Hempel and J. Luce (*Mitteilungen Kaiser Wilhelm Institut fur Eisenforschung*, 1941, **23,** *paper* 413, *translation* 303 *of Royal Aircraft Establishment*, 1949, March, pp.52) Tests at +20°, −78 and −188°C on three plain and three alloy steels are reported with an extensive literature review (42 refs).

Strain gauge calibration device for extreme temperature R. M. McClintock (*Rev. Sci. Instr.*, 1959, **30,** Aug., 715–718) A device for applying known strains to gauges is shown and measurements from 4–300°K are recorded.

Behaviour of round steel chains under static and fatigue stresses K. Wellinger and A. Stanger (*VDIZ*, 1959, **101,** Nov., 1425–1431) Results of tests on open-link round section chains buttwelded by two methods showed tensile strengths of 60–75% of that of the material. Condition of the material (tempered or normalized) had no influence on fatigue strength. The influence of link geometry and of chemical composition on tensile and fatigue strength were investigated.

Testing welded joints under bending A. S. Fal'kevich (*Lit. Proizv.*, 1959, (10), 41–42) Normal bending tests do not show a number of defects situated in depth. Such defects are more effectively revealed by tests 'up edge' for thicknesses between 15–20 mm. For welding joints welded on one side (for example the butts of tubes) the defects are best revealed by tests on the root of the joint.

Effect of shot blast and specimen size on the cold bending of plates T. Ohtake, H. Gondoh, Y. Fukuda, and A. Arima (*Tetsu to Hagane*, 1959, **45,** Sept., 988–991) Hardness distributions across the sheet, and other data, are given for rimmed, semi-killed, and killed steel sheets.

Elastic-plastic torsion of a circumferentially notched bar J. B. Walsh and A. C. Mackenzie (*J. Mech. Phys. Solids*, 1959, **7,** Oct., 247–257) This problem was first approached by developing mathematical theory in terms of a non-hardening continuum. Experiments were then carried out using pre-stretched Al alloy bars, and it was shown that the results agreed satisfactorily with the theory.

Alternating torsion tests on welded box sections R. P. Newman, T. R. Gurney, and G. Coates (*Brit. Weld. J.*, 1959, **6,** Nov., 534–541) This is the second interim report of an investigation of the fatigue strength of welded thin-gauge constructions. In the present report the results are described of alternating torsion fatigue tests on several thin-gauge box sections in mild steel made from pressings welded by the C-arc, metal-arc, and spot-welding processes. Such sections are common units of construction in the motor-car industry.—T.G.

The effect of heat-treatment and microstructure on the high-temperature ductility of 18%Cr 12%Ni 1%Nb steels K. J. Irvine, J. D. Murray, and F. B. Pickering (*JISI*, 1960, **196,** Oct., 166–179) [This issue].

On the hot-workability of round steel billets. II. Hot-torsion tests on segregation zones of the Cr–Mo steel billet Y. Nishio, T. Mimino, H. Kubotera, and H. Narita (*Tetsu to Hagane*, 1959, **45,** Sept., 1108–1110).

A study on the high-temperature torsion test of cast iron for ingot moulds K. Yoneda, A. Yamashita, T. Amaki, O. Tsuneda, and M. Fukuda (*Tetsu to Hagane*, 1959, **45,** Sept., 903–905).

Dynamic compressibility and equation of state for iron under high pressure L. V. Al'tshuler, K. K. Krupnikov, B. N. Ledenev, V. I. Zhuchikhin, and M. I. Brazhnik (*Zhurnal Eksp. Teoret. Fiz.*, 1958, **34,** (4), 874–885) Shock wave studies throughout the range 4×10^5–5×10^6 atm. are reported. The compressibility at 0°K is derived and the curve extrapolated to pressures at which statistical atomic models are valid.

The failure of the martensite crystal under the simultaneous action of external tension and compression stresses A. Mashin (*Fiz. Met. Metalloved.*, 1959, **8,** (6), 908–914) The investigations concerned the changes in the martensite crystal, substantially a monocrystal, under simultaneous external tensions and compressions. It was established that the martensite crystal, on failure, first breaks up, disintegrates, and then fractures. This interpretation was arrived at on the basis of an analysis of stresses acting on the crystal at the moment of failure. The disintegration of the crystal is explained by the fact that the compression stress on one side of the crystal is not compensated. The study was made in the course of investigating the distribution of the elastic and plastic waves in an austenite–martensite medium.

Supplementary study to the impact testing of cast iron A. de Sy and F. Deknock (*Fond. Belge*, 1959, **29,** (11–12), 266–271) Supplementing work reported in a previous article (*ibid.*, 1959, **29,** (10)), the authors describe the results of impact tests on specimens of diameters 10–20 mm, belonging to the proportional series L=5D÷1D. It is shown that it is necessary to operate with a ratio Er/Ed of at least 0·4, where Ed is total available energy, and Er=energy of rupture.

Impact resistance of nickel manganese cast steels I. B. Elman and R. D. Schelleng (*Mod. Castings*, 1960, **37,** Feb., 67–70) The effects of Ni and Mn, section size and deoxidation practice on 0·3%C cast steel were investigated. It was shown that Ni was essential in raising low temp. toughness. Of the compositions examined, 1·51%Ni, 1·6%Mn was found to be most suitable. Deoxidation with Si–Mn–Al was found to be better than Si–Mn alone. Larger sections showed a higher transition temp., but this effect was minimized by Ni.

Studies of load-time relations in the Charpy impact test. I. Effects of size of steel specimens on Charpy impact tests. II. On the transition temperatures of mild steels S. Sakui, T. Nakamura, S. Nunomura, and M. Omori (*Tetsu to Hagane*, 1959, **45,** Sept., 1082–1084; 1084–1085) I. (S. Sakui, T. Nakamura, and S. Nunomura). Absorption of energy in the test is discussed, and load-time curves are given for specimens of various thicknesses. II. (S. Sakui, T. Nakamura, and M. Omori). Graphical data are given for max. load and absorbed energy against testing temp. for normalized specimens.—K.E.J.

Study on low-manganese high-strength steel. I. Effect of various alloying elements on the notch-sensitivity of 2%Mn high-strength steel T. Saito and T. Fuziwara (*Tetsu to Hagane*, 1959, **45,** Nov., 1265–1272) Carbon increased notch sensitivity more than any other element. Si and Cr increase it in the sorbitic structure, but improve the impact value in the 300°C tempered range. Mo addition (0·5%) is always necessary. W and Ti were beneficial in the low tempering-temp. range. V was always beneficial, and Cu had no effect (12 refs).—K.E.J.

Some ductility aspects of 18-12-1Nb steel R. J. Truman and H. W. Kirkby (*JISI*, 1960, **196,** Oct., 180–188) [This issue].

Charpy-V transition temperatures of ferritic iron alloys: Part I. Iron–oxygen and iron–aluminium J. E. Srawley (*NRL Rep.* 5290, *PB* 151555, 1959, April 6, pp.17) The transition temp. of Fe increased steadily with O_2 content and, with the same O_2, was higher after furnace cooling after cold working than after quenching from 1650°F. Intergranular fracture was observed in annealed irons with 0·0039%O_2 or more and transgranular in quenched iron with <0·0206%O_2. Fe–Al alloys always show transgranular fracture and

the transition temp. passes through a maximum with increased Al content. **II. Iron–chromium and iron–chromium–aluminium** (*NRL Rept.* 5367; *PB* 151811; 1959, Aug. 24, pp.19; from *US Res. Rep.*, 1959, **32,** Dec. 11, 751) Alloys up to 24·3%Cr and three series with up to 25%Cr and 6%Al were examined after various heat treatments and conditions of mechanical working. Cooling conditions and embrittlement effects are discussed.

The low-temperature impact values of welded austenitic stainless steels M. Kawahata, K. Yokota, and K. Ebato (*Tetsu to Hagane*, 1959, **45,** Sept., 1013–1015) Charpy values are shown for the deposited and heat-affected zones at room and sub-zero temp. for six (AISI or NAS type) steels.—K.E.J.

Correlation of toughness between H-plate and Charpy-impact tests W. P. Hatch jun. (*Weld. J.*, 1960, **39,** Jan., 20s–30s) A series of laboratory tests were carried out Ni–Cr, Mn–Mo, and Ni–Cr–Mo armour steels, welded with several different electrodes, in order to try to correlate the performance of these armour steels in ballistic trials with the composite Charpy results. It was concluded that this was possible.

Creep, mechanical and metallurgical properties of an experimental Mo–V steel rotor forging A. I. Smith, E. A. Jenkinson, D. J. Armstrong, and L. M. T. Hopkin (*JISI*, 1960, **196,** Oct., 117–130) [This issue].

Heat-affected zone cracking in welded high-temperature austenitic steels R. N. Younger and R. G. Baker (*JISI*, 1960, **196,** Oct., 188–194) [This issue].

Temperature dependence of the paramagnetic susceptibility of nickel–zinc ferrites V. I. Chechernikov and Yu. D. Volkov (*Zhur. Eksp. Teoret. Fiz.*, 1958, **35,** (4), 875–879).

Temperature stability of the domain structure in iron silicate crystals L. V. Kirenskii and I. F. Degtyarev (*Zhur. Eksp. Teoret. Fiz.*, 1958, **35,** (3), 584–586).

The antiferromagnetism of γ-phase iron E. I. Kondorskii and V. L. Sedov (*Zhur. Eksp. Teoret. Fiz.*, 1958, **35,** (6), 1579).

Deterioration in the magnetic properties ('spoiling') of tungsten steel for F-233 magnets J. T. Martorell and J. M. Bermudez de Castro y Mosquera (*Inst. Hierro Acero*, 1959, **12,** 46–64) It is concluded that the deterioration observed is due to the presence of WC, which is formed principally during forging or rolling.

Structural changes and anomalous temperature dependence of the magnetic properties of Ni–Fe (50%Ni) alloy M. V. Dekhtyar and N. M. Kazantseva (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 412–416) The magnetic properties of this alloy quenched from 1200° show anomalies. On heating this alloy, its magnetic properties change rapidly within a narrow range (300–360°). The coercive force increases and the maximum magnetic susceptibility increases by a small amount. Magnetic saturation also increases, but on specimens tempered 100 h near 480° this anomaly cannot be observed. It is suggested that when heating this rapidly cooling alloy near 300° a process analogous to ordering begins in the disordered alloy. This is shown by the increase in the magnetic saturation and rise in the Curie point.

The formation of the cubic texture in the iron–silicon alloy Chou Pen-sin, Van Vei-min, and Chen Hen-chuan (*Fiz. Met. Metalloved.*, 1959, **8,** (6), 885–891) The fact that the cubic texture may be obtained in silicon–iron either with the primary or with the secondary recrystallization means that the production of high quality cubic textured transformer sheet steel can be reliable and that it is not necessary to use transversal rolling. It also means that the cubic texture which may be obtained in a large number of face-centred metal lattices can also be obtained in the body-centred lattice which is of great scientific value. It has been shown that no foci form for the cubic texture in the secondary recrustallization but that these already exist at the end of the primary recrystallization with the orientation (100) [001] and are surrounded by grains with the orientation (110) [001].

The thermomagnetic Nernst–Ettingshausen effect in nickel and iron–nickel alloys as a function of temperature R. P. Ivanova (*Fiz. Met. Metalloved.*, 1959, **8,** (6), 851–856) The measurements were carried out over a wide range of temp. and relations between this effect and the magnetic field, temp. and chemical composition of the alloy are given graphically.

On the component factors of specific losses and on the magnetic permeability of dynamo steel alloyed with phosphorus S. I. Doroshek and V. V. Druzhinin (*Fiz. Met. Metalloved.*, 1959, **8,** (6), 892–895) Small phosphorus additions to alloy steels result in a reduction of losses as shown by the hysteresis curves because of a coarsening of grains, resulting in a reduction of eddy current and in an increase in the electrical resistance. In a coarse-grained dynamo steel (10–20 grains per mm²) an increase in the P content results in an appreciable reduction of such losses. The initial and the maximum permeability also increases under the influence of P additions.

The anomalous behaviour of the coercive force in quenched steels tempered at high heat M. N. Mikheev and G. S. Tomilov (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 346–348) This is an attempt at explaining the anomaly in the coercive force of high heat-tempered martensite steels on the basis of Kondorski's theory (*Dokl. AN SSSR*, 1948, **63,** (5); 1949, **68,** (1)). By considering relatively small and relatively large magnetic inclusions it is confirmed that this anomaly in the coercive force is due to the change in structure, to the magnetic saturation and to the average dimensions of the carbide inclusions.

The Hall effect in iron–nickel alloys near the Curie temperature N. V. Bazhanova (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 342–345) Fe–Ni alloys with from 30·5 to 40%Ni by weight were investigated. With magnetic fields between 0 and 2000 Oe no effect of the field on the Hall emf could be observed in the ferromagnetic body. The method of measuring magnetic saturation and the emf has been described (*ZhETF*, 1957, **33,** 567). Findings for the Hall emf related to magnetic fields at various temp. near the Curie points are tabulated.

Hot-rolled transformer steel with low specific losses N. F. Dubrov (*Fiz. Met. Metalloved.*, 1959, **8,** (5), 752–757) The Verkh Isetsk Metallurgical Works has produced since 1958 on an industrial scale, hot rolled transformer steel in 0·35 mm thick sheets with a specific losses $P_{10/50}$ between 0·75 and 0·85 W/kg in tests with whole sheets. The characteristics of this steel are: coarse ferrite grains (6–10 grains/mm²) mainly with straight line boundaries which are mostly free from foreign substances. Small amounts of non-metallic inclusions, small amounts of Al_2O_3, S, N, and Al.

New eddy current test for tubing B. B. Burd (*Met. Prog.*, 1960, **77,** April, 101–103) The equipment, known as the 'Damascope' incorporates the 'Radac' multiple frequency electromagnetic unit developed by Metrol, Inc., background interference is eliminated. The equipment was used successfully for the location of surface and sub-surface defects in over 1000000 ft of tubing before being used in regular production.

Electrical conductivity of iron at high temperatures V. S. Gumenyuk and V. V. Lebedev (*Fiz. Met. Metalloved.*, 1959, **8,** (6), 847–850) Cylinders 3–6 mm dia. and 50–100 mm long were used as specimens for measurements by compensation by means of a potentiometer and galvanometer. The current source was an accumulator battery under continuous charge from a rectifier. A barretter was inserted in series in the circuit to maintain the current at 1 amp. The data giving the electrical resistance of the specimens with temp. from 20–1450° are tabulated.

Electrical resistance of iron, nickel and nickel–copper alloys at low temperatures E. I. Kondorskii, O. S. Galkina, and L. A. Chernikova (*Zhur. Eksp. Teoret. Fiz.*, 1958, **34,** (5), 1070–1076) The range 2–78°K is covered and the results are represented by expressions and discussed.

Electrical conductivity and the Hall effect at high temperatures, on iron alloys with silicon-containing leboite A. I. Gol'dberg, V. A. Lipatova, and P. V. Gel'd (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 472–475) The experimental findings confirm that conductivity in specimens with 40–45%Si is considerably greater than in others, and that the smallest conductivity is observed in alloys containing 49–51%Si.

Investigating the migration of iron tungsten alloy components in a constant electric field D. F. Kalinovich, I. I. Kovenskii, M. D. Smolin and I. I. Frantsevich (*Izvest. AN Metallurg. Topl.*, 1959, (**1**), 71–74) Data have been tabulated for iron and W of the variations in the speed of migration in the active zone, of the transfer numbers and of the charges on the iron atoms as a function of temp. Between 900–1000° all vacancies in the 3 d level of the iron atoms taking part in the electrical transfer are filled. At higher temp. a proportion becomes vacant and at 1150° all the 3 d levels are vacant. Thus between 900–1100° the valency electrons given out by the W atoms are filling the vacancies of the 3 d level of the iron atoms in a donor-acceptor relation.

Magnetic-field dependence of the ultrasonic attenuation in metals M. H. Cohen, M. J. Harrison, and W. A. Harrison (*Inst. Study Met. Univ. Chicago*, 1959, Dec., 408).

Ultrasonic weld inspection at work J. E. Bobbin (*Weld. J.*, 1960, **39,** Jan., 34–38) X-Ray and ultrasonic inspection methods are briefly compared, the speed, accuracy, and depth of penetration of ultrasonic waves being emphasized, and then examples of the use of ultrasonic weld testing in practice are given.

Hints on the radiographic investigation of castings R. Matthes (*Giessereitechn.*, 1959, **5,** Sept., 264–268) Technical considerations on the capacity and efficient use of X-ray sources are followed by a description of the use of a wire grid, and a discussion of the optical limitations of the method.

Some aspects of the radiography of thick plate welds E. H. S. Van Someren (*Welder*, 1959, **28,** July–Sept., 53–55) A review of X-ray and γ-ray methods.

X-Ray measurement of the atomic scattering factor of iron B. W. Batterman (*Phys. Rev.*, 1959, **115,** July 1, 81–86).

The influence of chromium diffusion upon the thermal oxidation properties of steel and of nickel- and cobalt-base alloys R. L. Samuel and T. P. Hoar (*Metallurgia*, 1959, **60,** Sept., 75–80) The results of experiments on the rate of oxidation of chromized steel and other materials are reported for various temperatures. It is shown that plain C and low-alloy steels can be effectively protected and given a high degree of oxidation-resistance at elevated temp. It is shown that the amount of oxidation in a creep-resisting chromized austenitic steel was about 1/6 of that of the untreated material, when both were oxidized in air for 400 h at 950°C; secondary diffusion was negligible.

Permeation of hydrogen through metals R. Ash and R. M. Barrer (*Phil. Mag.*, 1959, **4,** Nov., 1197–1206) Theory is developed mathematically.

Study of high-manganese steel. II. Effect of chromium on the heat-treatment characteristics of high-manganese steel A. Suzuki, H. Takada, and N. Kusuoka (*Tetsu to Hagane*, 1959, **45,** Sept., 1102–1104).—K.E.J.

On the function of nitrogen as an alloying element in heat-resisting materials. VIII. Effect of working on heat-resisting properties of 316L type steels and 16–15–6 type alloys M. Okamoto, R. Tanaka, A. Sato, and S. Aoki (*Tetsu to Hagane*, 1959, **45,** Sept., 1025–1027).

On the function of nitrogen as an alloying element in heat-resisting materials. IX. Effect of Mn content on the nitrogen absorption of 20%Cr–Fe Alloys M. Okamoto and O. Miyagawa (*Tetsu to Hagane*, 1959, **45,** Sept., 1020–1022).—K.E.J.

On the behaviour of nitrogen in ball-bearing steel M. Ohsawa (*Tetsu to Hagane*, 1959, **45,** Sept., 999–1000) The N_2 contents are examined in the raw materials used for steelmaking, the heat-treated steel, and the insol. residue after HCl extraction.—K.E.J.

Softening of steels under the action of hydrogen at high temperature N. N. Kolgatin, L. A. Glinkman, V. P. Teodorovich, and V. I. Deryabina (*Metalloved., Obra. Met.*, 1959, (7), 16–21) Under the action of H_2 at high temp. and pressures, decarburization occurs in such steels as 20, 30KhMA, and EI 579. Decarburization is considerably more intense in

specimens under pressure from all sides than in those under pressure on one side only (internal) Gradual appearance of cracks is observed in the decarburized zone along the grain boundaries, which reduce considerably the intercrystalline strength.

The effect of additions of nitrogen and boron on the properties of heat-resisting steels of the CrNiFe(30:30:30) type J. Chodorowski (*Hutn. Listy*, 1959, **14,** (12), 1141–1143) Up to 0·4%N increases the creep strength and creep life considerably in these alloys, which also contain small amounts of Mo and Nb. B is less effective in this respect, but in contrast to N_2, has a favourable effect on ductility. The use of both elements results in pronounced improvement in creep properties and in the structure of the alloys in the as-cast state.—P.F.

The effect of the composition and of the method of deoxidation of steel on the liquation of phosphorus O. D. Moldavskii (*Izvest. AN Metallurg. i Topl.*, 1959, (5), 126–127) An increase in the carbon content affects the properties of the steel during solidification in the same way as Si. C displaces P from its solution in iron, increasing the degree of liquation and increasing the P content in the non-metallic inclusions. Deoxidation of the steel by Al, on the contrary, results in a reduction of liquation.

Effect of phosphorus on structure transformation and properties of low alloy electrotechnical steel S. I. Doroshek (*Fiz. Met. Metalloved.*, 1959, **8,** (5), 770–776) An increase of the phosphorus content in low-alloy Si steel promotes grain growth and raises recrystallization temp. On structure, the effect of P is indicated by an improvement of the magnetic characteristics of the steel.

The effect of the inoculation additions on the mechanical properties of cast alloy steel M. P. Braun (*Izvest. AN Metallurg. i Topl.*, 1959, (4), 59–65) Compositions of the steels investigated were 0·25%C, 1·26%Cr, 1·42%Si; 1·09%Mn, 1·14%Ni, and 0·25%C; 1·22%Cr, 0·98%Si; 1·33%Ni. Inoculations affect these steels variously. The greatest increase in the UTS and yield strength is obtained with Se in Cr–Si–Mn–Ni and in Cr–Si–Ni steels. In a Cr–Mn–Ni steel with 0·5%Si, Se increases the plastic properties. V addition to the third mentioned type of steel also improves its properties while it has no such effect on the other two types. The harmful effect of Ti and Nb inoculations are explained by the fact that these elements pass entirely into carbides and do not dissolve in the α-phase because of their low solubility.

Effect of titanium on Cr–Ni austenitic steels R. Nakagawa and Y. Otoguro (*Tetsu to Hagane*, 1959, **45,** Sept., 1004–1006) The effects on ageing hardness, high-temp. tensile strength, and creep-rupture properties are examined.—K.E.J.

The effect of niobium on the high-temperature properties of austenitic steels R. Nakagawa, and Y. Otoguro, and T. Tsuchiya (*Tetsu to Hagane*, 1959, **45,** Nov., 1276–1281) The specimens were of 18–12 Cr–Ni stainless steels. Creep-rupture strength increased with Nb content, but became constant beyond Nb= 11× (C content). Nb had no effect on high-temp. tensile strength. High-temp. deformation resistance depended only on the Nb dissolved in the matrix.—K.E.J.

Effect of Mo, W, V, and Nb on properties of 38%Ni heat-resisting alloy S. Koshiba and T. Kunou (*Tetsu to Hagane*, 1959, **45,** Sept., 1031–1032).—K.E.J.

Effects of W, Mo, and V on the various properties of high-C, high-Cr steel (8KDI) S. Koshida and K. Kiyonaga (*Tetsu to Hagane*, 1959, **45,** Sept., 1104–1106) Data are given for the effects on oil-quenched hardness, tempering hardness, change of torsion ratio by quenching temp., and abrasion properties.

Study of low-alloy structural steel. I. Effect of alloy elements on the hardenability of low-alloy structural steel (SNCM8). II. Effect of alloy elements on the temperability of low-alloy structural steel (SNCM8) T. Nakano, M. Makioka, and M. Maeda (*Tetsu to Hagane*, 1959, **45,** Sept., 1111–1113) **I.** Graphical data are given for effects of varying Mo, Cr, Ni, (Mo+C), and (Mo+V) on the quenching hardness. **II.** T. Nakano, M. Makioka, and E. Niina (1113–1115) The effects of varying the same elements on the resistance to tempering and the impact value are shown.—K.E.J.

Study on high-strength structural steel. I. Effects of addition of V and other elements on Si–Mn steel E. Asano and A. Suseki (*Tetsu to Hagane*, 1959, **45,** Oct., 1158–1163) Increase of V increases tensile and yield strengths, unifies microstructure, refines ferrite grain size, and reduces softening by tempering. In Si–Mn–V steel, Cr increases tensile strength, and Ti and Zr refine the ferrite grain size. Si (0·4–0·65%) has no effect on tensile strength, even with C contents as low as 0·1%. **II. Effect of B and Mo on Si–Mn and Si–Mn–Cr high-strength steel** E. Asano and T. Ishida (Sept., 1110–1111) Data are given for hardness changes of Mn–Si, Mn–Si–Cr, Mn–Si–Mo, and Mn–Si–Cr–Mo steels, as rolled, normalized, quenched, and tempered, and for Jominy hardnesses up to 95 mm from the quenched end.—K.E.J.

Influence of small quantities of nickel and chromium on the mechanical properties of steel T. Ohtake, S. Murayama, and N. Eguchi (*Tetsu to Hagane*, 1959, **45,** Sept., 1089–1092) Results are given for the effects of up to 0·5% of the combined elements on tensile and impact properties and transition temp. of 0·1 and 0·6%C steels.—K.E.J.

Effect of various elements on the properties of 25% chromium–12%nickel steel F. Tsukamoto, H. Sunaga, and T. Suzuki (*Tetsu to Hagane*, 1959, **45,** Sept., 1027–1029) The effects of C, Mn, Si, Cr, and Ni on tensile, hardness, impact, and creep-rupture properties are shown.—K.E.J.

Relations between the mechanical properties of cast steels and their analysis. Investigation of practical formulas J. Jamar (*Fond. Belge*, 1959, **29,** (11/12), 272–274) The work refers to hypoeutectoid steels, normalized and tempered, and the formulas derived take into account (1) C only, with Mn between 0·7 and 0·9%; (2) C and Mn. A note on the influence of other elements is added.

The development of heat resisting austenitic utility steels in Czechoslovakia F. Pobořil (*Hutn. Listy*, 1959, **14,** (12), 1121–1125) The development of a number of steels in which N is replaced partly or entirely by other alloying elements, such as Mn, N_2, or Mo with V, is discussed and their composition and properties are surveyed. Seamless tube production, argon-arc welding, and other associated processes are considered.—P.F.

METALLOGRAPHY

The interpretation of contrasts in electron micrographs obtained by transmission of thin crystals H. Alexander (*Z. Metallkde.*, 1960, **51,** April, 202–209) Thin metal films (100–1000 Å thick) give many bright-dark contrasts when submitted to transmission electron microscopy. The use of the diffraction theory of the electron waves leads to an interpretation that part of these contrasts are diffraction images of linear or two-dimensional lattice imperfections such as vacancies or stacking faults.—T.G.

Specimen holder for deforming whiskers in an electron microscope F. R. Rollins jun. (*Rev. Sci. Instr.*, 1959, **30,** July, 603–604).

On the new polishing method for the development of the micro-structure of steel. I. A polishing method for carbon steels and low-alloy steels and its results K. Kawamura and R. Kazui (*Tetsu to Hagane*, 1959, **45,** Sept., 1053–1055) The new method uses optimum H_2O_2–$C_2H_2O_4$–H_2O and H_2O_2–HF–H_2O solutions.—K.E.J.

The nature of mechanically polished surfaces (*Annual report of the Defence Standards Laboratory, Commonwealth of Australia, for the year ended June* 1959, 26–28) The work done recently has been on industrial polishing. The findings are that polishing is a fine cutting process, and that a Beilby layer does not occur. The plastic deformation in the surface layers is largely compressive in nature, and, when very severe, may cause the collapse of internal cavities exposed at the surface, and the fragmentation of non-metallic inclusions. **Radio tracer studies of reactions occurring at metal surfaces** (28–29) The reaction between zinc and corrosion—inhibiting chromate solutions is being studied, and the conclusions reached so far are presented.

Improved method of etching by ion bombardment T. K. Bierlein and B. Mastel (*Rev. Sci. Instr.*, 1959, **30,** Sept., 832–833) An improved etching chamber of Pyrex is shown.

The isolation of carbides from steels and similar alloys H. Tûma (*Chem. Listy*, 1959, **53,** (9), 909–921) The chemical and electrolytic methods of separating carbides from steels, the separation of the carbides according to composition, as well as recent methods developed for their analysis are discussed with reference to publications which have appeared in the last few years.—P.F.

A new optical technique for the determination of non-metallic inclusions in steel T. Ernst and G. König (*Stahl Eisen*, 1960, **80,** April 28, 626) The authors describe briefly a new technique for determining directly refractive index and true colour of transparent non-metallic inclusions in metallographic sections of steel under oblique illumination. The novelty is to remove metal by electropolishing, thus leaving the non-metallic inclusions slightly protruding above the metal surface (11 refs).—T.G.

Inclusions help identify stainless steels R. P. Wentworth (*Met. Prog.*, 1960, **77,** March, 125–126) The use of microscopical examination to distinguish between types 321 and 347 (stabilized austenitic) and 304 (unstabilized) stainless steels is described.

Studies on the origin of non-metallic substances in steel ingots T. Ao and T. Tokuda (*Tetsu to Hagane*, 1959, **45,** Sept., 1043–1045) A scheme is set out showing the composition and possible origins of all inclusions, surface defects, etc., throughout the ingot.—K.E.J.

Deoxidation and inclusions in the semi-killed steel ingot T. Kato, J. Imai, and H. Kajioka (*Tetsu to Hagane*, 1959, **45,** Sept., 1042–1043).

Study of non-metallic inclusions in iron and steel with the electron microscope. I. On titanium carbide K. Narita and T. Mori (*Tetsu to Hagane*, 1959, **45,** Sept., 1052–1053) Electron micrograph and electron diffraction data are given and analysed.—K.E.J.

On the study of non-metallic inclusions in steel. I. Effect of non-metallic inclusions on the mechanical properties of ball-bearing steel I. Uchiyama and M. Ueno (*Tetsu to Hagane*, 1959, **45,** Sept., 996–998) The occurrence of inclusions and the mechanical properties are correlated for steels (analyses given) deoxidized by Al, Si, or Mn.—K.E.J.

Study of non-metallic inclusions in steels. I. On the composition of non-metallic inclusions in trial ingots by the Klinger–Koch method Y. Shimokawa, T. Fujii, and T. Yamamoto (*Tetsu to Hagane*, 1959, **45,** Sept., 1047–1049).

The study of non-metallic inclusions. VI. On the deoxidation products due to Si, Mn and Al M. Kawai, H. Kimura, and T. Noda (*Tetsu to Hagane*, 1959, **45,** Sept., 1038–1041) Data are given for melting practice, cleanliness, and sand marks.—K.E.J.

Identification of non-metallic inclusions consisting of the sand mark by means of electron diffraction K. Narita and K. Mori (*Tetsu to Hagane*, 1959, **45,** Sept., 1045–1047) Data are given for α-Al_2O_3, FeO–Al_2O_3, α-Fe, and substances 'A' and 'B'.—K.E.J.

On the determination of non-metallic inclusions in low-chromium steels by chlorine methods Lee Tai-chung, Chao Shu-hsi, and Lu Tsin-hua (*Acta Met. Sin.*, 1959, June, **4,** 162–168) Inclusions in steels having >0·5%C can be quant. determined by a combination of the normal-pressure Cl_2 and electrolytic methods. The low-pressure Cl_2 method can be adapted for ball-bearing steels. The practicability of the $FeSO_4$ electrolytic method is confirmed (14 refs).—K.E.J.

Non-metallic inclusions in stainless steels (18–8–Ti and 18–8–Nb steel) T. Tanoue and S. Inoue (*Tetsu to Hagane*, 1959, **45,** Sept., 1002–1004) The quantities of inclusions are related graphically to the contents of C and N.

On carbides in 12% chromium heat-resisting steels. I. Effect of Ti and Nb on the behaviour of carbides in 12% Cr steel T. Fujita, M. Tanino, N. Sano, and S. Masumoto (*Tetsu to Hagane*, 1959, **45,** Sept., 1018–1020) Results of X-ray and micrographic tests on the carbides, and

data on the change of composition of the carbides with temp. and the hardness change with tempering are given.—K.E.J.

The effect of Mn-containing carbide on the determination of unstable non-metallic inclusions in steel Lee Chin-yuan and Chao Shu-hsi (*Acta Met. Sin.*, 1959, Sept., **4,** 282–286).

An investigation of ghost lines in a low-alloy steel Ho Tsen-chian, Chen Chi-tse, Fung Munchu, and Chu Yuan-tei (*Acta Met. Sin.*, 1959, June, **4,** 114–126) Billets of 30 Cr–Mn–Si steel made in a basic arc furnace were investigated. The smaller the dia. the longer the ghost lines. The lines usually originated from chain-type inclusions such as spinel. They can be reduced or eliminated by improved deoxidation, an optimum degree of forging, and by removing the surface layers of ingots.—K.E.J.

The intermediate carbide phases in carbon steels N. V. Gudkova and V. P. Kuznetsova (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 468–470) In view of the data obtained by Apaev (*ibid.*, 1957, (1)) the question of differences in the phase composition of annealed carbon steels in relation to their carbon content became important. Since these differences cannot be observed by magnetic methods in steels containing less than 0·4%C the specimens were investigated by diffraction and the relevant maps obtained are tabulated.

Chromium distribution around grain boundary carbides found in austenitic stainless steel B. E. Hopkinson and K. G. Carroll (*Nature*, 1959, **184,** Nov. 7, 1479–1480) A letter. Using the microprove analyser a 25Cr–20–Ni stainless steel with 0·15%C was variously heat treated. Cr depletion was not observed. The thickness of the depleted area if present must be $<1\,\mu$.

The nature of phases formed in the structure of brittle transformer sheet J. Ježek, J. Vobořil, and V. Číhal (*Hutn. Listy*, 1959, **14,** (9), 777–786) It is shown that brittle nitride phases can be dissolved in the ferritic matrix by annealing above 800°C followed by cooling (200°C/h). (41 refs).

Crystallization of a solid solution M. V. Pikunov (*Zhurn. Fiz. Khim.*, 1959, (10), 2253–2258) The crystallization equilibrium is considered as the sum of two processes: precipitation of the solid along the solidus and the interaction between the precipitate and the liquid phase. Tests here described have shown that the interaction is shown in the transition of the high-melting component (raising of the liquidus point) from crystals to liquidus together with a transition of the low melting component (lowering of the liquidus point) from crystals to liquid. The mean composition of the component passing from liquid to crystals has been plotted on a curve.

Influence of carbide-forming elements on grain size determination by the carburizing method A. Adachi and K. Mizukawa (*Tetsu to Hagane*, 1959, **45,** Sept., 1057–1059) The influence of Cr and Mo in 14 steels (analyses given) is studied and the carburizing and quenching methods are compared.—K.E.J.

Eutectoid crystallisation E. Scheil and A. H. Paksad (*Z. Metallk.*, 1959, **50,** Oct., 582–587) It is shown by investigations on steel and nonferrous alloys that laws governing eutectic crystallization are to some extent valid for eutectoid crystallization. Differences arise mainly from the fact that in the former case crystallization occurs from a liquid, and in the latter from existing crystals. Differences are discussed with reference to the formation of pearlite in steel.

On austenite grain growth in stainless steels T. Mori and E. Ichise (*Tetsu to Hagane*, 1959, **45,** Sept., 1006–1008) The influence of holding time (up to 1 000 h) on grain growth at temp. between 800° and 1 100°C is shown graphically.

Origin, structure and some properties of the crystal boundaries of cast metals and alloys B. A. Movachan (*Izvest. AN Metallurg. i Topl.*, 1959, (5), 77–85) Investigations are described of some special features of plastic deformation under tension, within the grain boundaries of technical iron. These were carried out by interferometry, optical and electron metallography. In cast metals and alloys of the solid solution type, the new crystals forming during polygonization, after full solidification, should be distinguished from the others. Their boundaries have a great effect on the tendency of the cast alloys to the different types of intercrystalline failure because they are microsections with a great density of incomplete structure crystallization and therefore with the highest probability of microcrack formation and because during cooling, they may become enriched by certain elements causing brittle intercrystalline failure.

The formation of cube texture in an iron silicon alloy Chou Pang-hsin, Wang Wien-min, and Chen Neng-kuan (*Acta Phys. Sin.*, 1960, **16,** (3), 155–159) Iron with 3·25%Si was used. Cubic texture can be developed by either primary or secondary crystallization processes and conditions are indicated. Magnetic torque measurements show the amount of (100) [001] texture to be >90%.

The importance of the structure of iron phases in the crystallisation of cast iron E. Scheil (*Giess. Techn. Wiss.*, 1959, (26), Oct., 1455–1458) The structure of graphite is discussed, and the behaviour of γ-solid solution, the liquid and cementite towards graphite is reviewed. Mechanism of formation of C_2 molecules as a preliminary stage in the nucleation of graphite is discussed, and the greater difficulty of nucleation from γ-solid solution compared with the melt is shown theoretically, and verified by reference to published results.

Contributions to the study of austenite recrystallization in overheated steel B. Rotenstein and N. Dragan (*Rev. Roumaine Met.*, 1959, **4,** (2), 131–143) [In English] Experimental findings have shown that recrystallization in previously overheated austenite occurs also with moderate rates of heating (150–200°C/min) used for the second austenitization. Recrystallized austenite with a fine grain preserves this grain size over a wide range of temp. between 900 and 1 050°. No connexion was found between the modification of the austenite grain size as a function of the usual heating temp and that as a function of the temp. at which the austenitizing of the overheated steel was carried out before the second quenching.

Rolling and recrystallisation texture of transformer steel H. Möller and H. Stäblein (*Techn. Mitt. Krupp.*, 1959, **17,** (3), 162–164) The radiographic method used is described, which can be applied to the determination of crystal orientation on coarse-grained specimens.

Metallographic study of cold-rolled low-carbon steel sheets. I. Fragmentation of grains in cold-rolled low-carbon steel sheets. II. Geometrical study on ferrite grains and pearlite colonies in cold-rolled low-carbon steel sheets M. Osawa (*Nippon Kinzoku*, 1959, **23,** March, 145–149; 149–152) I. Comparison of the microscopical and external appearances of sheets shows that pearlite colonies and ferrite are split into smaller grains by cold-rolling, the grains more so than the colonies (12 refs). II. The observed mean distances between colonies, and the degree of elongation of colonies and grains are much different from estimates made from the change of dimensions of the specimens. There is a linear relationship between the logarithms of degree of elongation and the hardness of ferrite and pearlite.

The effect of the duration of the treatment on the phase composition of iron electrically hardened by graphite B. P. Kolesnik (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 470–472) Electric spark hardening of components by graphite affects the phase composition of the surface layer up to a duration of 6 min/cm² for a given treatment during which time the phase composition and therefore the properties of the surface layer change within wide limits. The hardening of components and instruments takes place between 0·5 and 2·5 min/cm², i. e. in the field of the maximum effect of the whole period of treatment due to phase composition. If therefore the time is not rigorously controlled, different results may be obtained.

On the polygonized structure of iron-aluminium alloy B. V. Molotilov (*Fiz. Met. Metalloved.*, 1959, **8,** (3), 463–465) The material investigated is a high-permeability alloy of iron with 16%Al. After slow cooling between 980 and 590° and rapid quenching at 590° the alloy has a maximum permeability of 71 000 Gauss/Oersted. The structure was examined after electrolytic polishing in H_3PO_4 and electrical etching. The structure was also observed by Lacombe and Beaujard (*J. Inst. Met.*, 1948, **74**) and by Cahn (*idem*, 1949, **76**) in Al and by Dunn and Hubbard (*Acta Met.*, 1955, **3**) in Fe–Si alloys and by other investigators in nonmetallic crystals. The formation of such a structure during the anneal after deformation may be connected with polygonization resulting in the formation of ordered systems of dislocation caused by their displacement in the crystal lattice at the annealing temp.

Effect of hot working conditions on the microstructure and mechanical properties of carbon steels Chang Tso-mei and Li Chien-hwa (*Acta Met. Sin.*, 1959, June, **4,** 143–154) Large austenitic grain size leads to non-uniform long grains, and banded and Widmanstätten structures. Increase in this grain size lowers the yield pt. and impact values at −20°C. Increase in cooling speed after rolling raises these parameters and room temp. and −20°C. For steels with different carbon content rolled at 1 050°C, there is a critical reduction range between 0 and 10% for coarsening of austenitic grain size. Reduction in the last hot-working operation has the greatest effect.—K.E.J.

Behaviour of carbide and nitride in the transformation of carbon steels C. Asada, S. Nagata, and K. Kato (*Tetsu to Hagane*, 1959, **45,** Sept., 1059–1061).—K.E.J.

Phase transformation in carbon steels subjected to high rates of heating G. T. Fomin (*Metalloved. Term. Obra. Met.*, 1959, (8), 54–55) The experimental apparatus was described in *Zavods. Lab.*, 1958, (1), by the author. U8A (0·83%C, lamellar pearlite structure) and U10 (0·99%C, pearlitic grain-structure) steels were investigated. Curves have been plotted from the readings which show that the transformation takes place in three stages; the polymorphous transformation of ferrite into low carbon austenite; the partial transformation of cementite to low carbon austenite (both these processes proceed to completion of the polymorphic transformation of the ferrite) and the final solution of the cementite in austenite.

Phase transformation in the iron–vanadium system M. I. Sakharova, L. A. Semenova, and F. N. Stetsenko (*Izvest. AN Metallurg. i Topl.*, 1959, (5), 135–138) Structure and magnetic properties of the Fe–V alloys have been investigated and as a result the existence of a second high temp. magnetic phase (β-phase) has been established, with a Curie point at about 200°C, crystallizing in the face-centred cubic lattice.

The kinetics of the martensite transformation in iron alloys W. S. Owen and A. Gilbert (*JISI*, 1960, **196,** Oct., 142–149) [This issue].

Isothermal transformation diagrams for some Swedish standardized heat-treatable steels G. Molinder (**Jernkont. Tekn. Rdd.*, 1959, **20,** (255), 405–430) Steels SIS 1, 1672, 2120, 2225, 2234, 2244, 2530, 2534, 2536, 2541, and 2940 were tested in several laboratories and the agreed *TTT* diagrams are shown.

Isothermal transformation diagrams for four Swedish standardized spring steels G. Molinder (**Jernkont. Tekn. Rdd.*, 1958, **20,** (248), 155–164) Steels SIS 14, 1770, 2090, 2230, and 2231 are reported on. Compositions and heat treatments are indicated and the diagrams are shown.

Differences in the process of carbide formation at the boundaries and within the grains L. G. Orlov and L. M. Utevskii (*Izvest. AN Metallurg. i Topl.*, 1959, (3), 132–134) Electron microscope investigations have shown that in spite of the fact that for the growth of the particle of special carbides a diffusion supply of carbon and chromium is necessary, the conditions at the grain boundaries are so different from those inside the grains that even a speeded up growth of particles there is ensured by a diffusion flow of chromium alone along the contiguous grains.

The nature of the toughening of quenched high speed steel in tempering P. M. Yushkevich (*Fiz. Met. Metalloved.*, 1959, **8,** (6), 896–903) The secondary hardening is not a result of the transformation of the residual austenite but is a consequence of the dispersion of the primary

and secondary martensite (dispersion hardening). The dispersion hardening due to the residual austenite is of secondary importance. The reduction in hardness during the initial period of tempering is accompanied by a non-diffusive expansion of the lattice of the residual austenite.

Heterogeneity of cementite in ball-bearing steel K. Seto (*Tetsu to Hagane*, 1959, **45**, Sept., 1001–1002).—K.E.J.

Factors influencing graphite or carbide formation in cast iron A. Bordes (*Metalen*, 1960, **15**, Feb. 15, 26–40; March 15, 78–80; March 31, 94–96) These instalments of this lengthy article on graphite formation in cast iron discuss the influence of cooling, the separation of graphite in austenite, the reabsorption of small graphite inclusions, and the reasons for the occurrence of the various kinds of graphite formation. Then the question of graphite formation in white, grey, and nodular cast irons is dealt with. In each instalment there appear numerous graphs and structures of the different types of cast iron concerned.—F.R.H.

The influence of titanium and nitrogen on graphitization of malleable cast iron Y. Ueda (*Imono*, 1959, **31**, Aug., 719–723) Ti forms a nitride until all N_2 is used up and then forms TiC. The Ti =N produces maximum effect so that probably the nitride produces the graphitization. Preheating of the white iron, when N more than equivalent to the Ti is present, is also effective.

The effect of C, Si, Mn and S on the second stage of graphitization of cast irons S. Drápal (*Hutn. Listy*, 1959, **14**, (12), 1064–1073) A detailed study was made of the graphitization kinetics in cast irons containing 2·3–3·3%C, 0·6–1·6%Si, 0·4–0·7%Mn, 0·08%P, 0·14%S, 0·06%Cr, and 0·016–0·030%Al. Carbon has no effect on the formation and growth of ferrite, which is effectively aided however by Si. Excess Mn and S retard the process. If S is not fully combined with Mn an anomalous decomposition of austenite is observed. The elements used as additions have a similar effect on the decomposition of pearlitic cementite as on the transformation of austenite.—P.F.

On carbide precipitation by reheating in solution-treated austenitic manganese steel and the effect of a small amount of added, V, Ti and Zr Y. Imai, T. Saito, and I. Nakazawa (*Tetsu to Hagane*, 1959, **45**, Sept., 1100–1102) Isothermal and continuous-heating pptn. diagrams are given.—K.E.J.

Transformation of austenite into martensite at high pressures A. I. Stregulin and L. A. Mel'nikov (*Fiz. Met. Metalloved.*, 1959, **8**, (3), 406–411) A special magnetometric apparatus is described to investigate this transformation at pressures up to 30 000 kg/cm². The findings show that pressure considerably reduces the initial temp. of the martensite transformation and decreases its effect. The isothermal austenite transformation into martensite occurs more slowly at high pressures than at atmospheric pressure.

CORROSION

Caustic cracking of steel pipe H. M. Wilten (*Met. Prog.*, 1960, **77**, March, 128, 130) A letter. Examination of a cracked weld in a pipe used in unloading caustic alkali at ambient temp. and a pressure not greater than 40 lb. showed main and minor cracks to be intercrystalline. No evidence of corrosion was found in the inside of the pipe.

An investigation relating to the corrosion of coke cars by spent ammoniacal liquor T. K. Ross, R. L. Cooper, and J. E. Meeks (*Coke Gas*, 1960, **22**, April, 143–146) Corrosion of mild steel was examined and chloride was shown to be the main corrosive agent.

Further experience of car radiator corrosion T. H. Turner (*Corros. Prev.*, 1959, **6**, Dec., 41) Three new standards published by BSI for acceptable varieties of corrosion-inhibited ethylene glycol anti-freeze for water-cooled engines are enumerated, and experience of the author is cited to show that it is advisable to keep a car radiator filled with such a solution all the year round to inhibit corrosion.

Corrosion problems in the petroleum industry (*Chem. Ind.*, 1960, Jan. 23, 92–93) Symposium held Nov. 26–27, 1959, by the Corrosion and Chemical Engineering groups of the S.C.I. and the Institute of Petroleum jointly. **Corrosion of production equipment and gathering lines and its prevention** G. A. Lee and G. A. Haines (92). **Corrosion of marine structures and its prevention** G. T. Colegate (92). **Corrosion of tanker hulls and its prevention** J. G. Robinson and K. Fleming (92). **Developments in the protection of storage tanks and pipelines** H. B. Footner and P. W. Heselgrave (92–93). **Corrosion problems arising in tubular vent-exchange equipment** P. T. Gilbert (93). **Corrosion control in reticulations using salt water** E. D. Dolan (93). **Non-destructive testing** L. Wilkinson and P. S. Cotton (93). **Developments in corrosion-resisting materials: metals** G. L. Swales (93). **Developments in corrosion-resistant materials: non-metallic materials** H. Thomas and T. S. Roberts (93).

Corrosion of furnace tubes by residual welding slag K. L. Moore (*Corrosion*, 1960, **16**, Jan., 26t–28t) Severe and rapid corrosion of hydrogen reforming furnace tubes in an oil refinery during high-temp. operation is reported. The attack was caused by the presence of residual welding slag and sulphur and by the reducing conditions prevailing. The best remedy against this type of corrosion was thorough sandblasting.—T.G.

Corrosion in salt water cooling systems E. D. Dolan (*Corr. Eng.*, 1959, Aug., ii–x, supp. to *Corr. Prev. Control*, 1959, **6**, Aug.) Corrosion of various parts of the system and cathodic protection methods are reviewed.

Accelerated oxidation by vanadium pentoxide L. Fairman (*Chem. Ind.*, 1959, Nov. 1, 1436–1441) A literature survey with a section on corrosion (31 refs).

Corrosion studies: calcining aqueous radioactive wastes E. J. Tuthill and R. F. Domish (*Ind. Eng. Chem.*, 1959, Dec., 1471–1474) Stainless steel and other high temp. alloys proved suitable for construction of the double-kiln calciner.

Corrosion problems in heterogeneous nuclear reactors V. Čihal and I. Sekerka (*Hutn. Listy*, 1959, **14**, (11), 978–984) Corrosion problems arising in reactors cooled by high-pressure CO_2 and water are surveyed, and the conditions under which certain alloys are most effective are considered. Stainless steels of the 18–8 type are suitable up to 360°C, but high Cr-steels (up to 31%Cr) with a low Ni content (2·5%) have a comparatively low corrosion resistance. Alloys of the Nimonic and Hastelloy type are also suitable for use in water up to about 350°C.—P.F.

The structure of the alloy layer in the reaction of iron–carbon alloys with liquid aluminium L. J. Hütter, H. H. Stadelmaier, A. C. Fraker, and R. G. Hardy (*Z. Metallk.*, 1959, **50**, Oct., 625–627) It is shown that the alloy layer formed by the attack of molten Al on white and grey cast iron consists essentially of Al_3Fe_2 and Al_4C_3; no trace of the ternary carbide Fe_3AlC_x was found by radiographic examination, but microscopical investigation indicates that small amounts of this phase probably occur in a narrow zone adjacent to the iron.

The kinetics of the reaction of solid and liquid aluminium with iron T. Heumann and S. Dittrich (*Z. Metallk.*, 1959, **50**, Oct., 617–625) Diffusion of Al in Fe is determined almost entirely by the behaviour of the η-phase Al_5Fe_2. Its preferential formation is due to an abnormally high diffusion rate, which in turn is attributable to the structure of the phase. A very high diffusion rate occurs in the direction of the C-axis of the phase, and the activation energy of the diffusion process is 13·1 kcal/g.atm.

Corrosion of carbon steel in ethylene dichloride G. P. Maitak, I. A. Gutsev, and N. S. Kozachenko (*Ukrain. Khim. Zhurn.*, 1959, **25**, (4), 525–531) The investigations studied the corrosion of carbon steel in $(CH_2Cl)_2$ vapour and liquid and in its solution in alcohol (3:1 by volume) at different degrees of purity of the solvent, over a range of humidity and temp. Corrosion increases with temp. and diminishes with air admission. It is considerably more intense in the vapours of the boiling solution than in the liquid. In sealed ampoules in which the volume of O_2 from the air was limited, corrosion was greater than in the liquid phase. After the tests the specimens exposed to the air rusted very rapidly. Alcohol and moisture increase corrosion appreciably, especially at higher temp. The presence of marble diminishes corrosion.

On the nature of the heat resistance of modified 12%Cr steels J. Koutský (*Hutn. Listy*, 1959, **14**, (11), 951–955) Metallographic, chemical, and mechanical tests were carried out on 12%Cr steels containing 0·13%C, 0·14–0·9%Ni, 0·8–3·9%W, 0·18–0·28%V, and 0·018–0·06%N, after various quenching and tempering treatments. The creep resistance is related to the precipitation processes observed, and the effect of small amounts of V and Mo on their course is discussed.—P.F.

Factors affecting the adherence of oxides on metals R. F. Tylecote (*JISI*, 1960, **196**, Oct., 135–141) [This issue].

Kinetic study of the oxidation of iron at high temperatures and under low oxygen pressures P. Belin (*Corros. Anticorros.*, 1960, **8**, March, 96–112) A mathematical review in three sections: (I) equipment and experimental methods; (II) preliminary tests and examination of secondary factors; (III) analytical study of the isotherm-isobars of oxidation. A new general method of analysis is described which is claimed to be applicable whatever the nature of the parabola followed. Armco and Puron irons were used.—S.H.-S.

Studies on the metallurgy of silicon iron. IV. Kinetics of selective oxidation A. U. Seybolt (*Trans. Met. Soc. AIME*, 1959, **215**, Oct., 756–759) The kinetics of the selective oxidation of a 3·2%Si–Fe alloy in H_2O–H_2 gas mixtures at the dew point were examined over the temperature range 800–1000°C. It was shown that the parabolic law was apparently observed in some temp. regions near each limit of the range studied, but a non-parabolic behaviour was noted at intermediate temp. It is considered that this may be due to a lowered Si activity, due to partial Fe–Si ordering in the alloy solid solution.

The oxidation of iron–nickel alloys: III. Kinetics of oxidation of three commercial alloys R. T. Foley and C. J. Guare (*J. Electrochem. Soc.*, 1959, **106**, Nov., 936–940) Alloys with 30, 41, and 78%Ni were oxidized at 600–1000°C in air and in a 21·7%O_2–78·3%N_2 mixture. A rate-determining process of cation diffusion through a spinel structure is apparently concerned.

The study of the passivity of metals in inhibitor solutions, using radioactive tracers. Part 2: The mechanism of film growth O. Kubaschewski and D. M. Brasher (*Trans. Farad. Soc.*, 1959, **55**, July, 1200–1206) A theory of the log. growth of oxide films in which electron transfer through the film is the rate-controlling step.

On the passivity of iron in hot concentrated alkaline solutions A. M. Sukhotin and K. M. Kartashova (*Zhurn. Fiz. Khim.*, 1958, **32**, (9), 2068–2072) The passivating film of iron in a 10 N NaOH solution at 80° as well as in acid solutions consists of Fe_3O_4. The passivating action is due to a very thin surface skin, although the film may attain appreciable thickness. The characteristic features are determined by the electrochemical properties of the film.

Atmospheric factors affecting the corrosion of steel P. J. Sereda (*Ind. Eng. Chem.*, 1960, **52**, Feb., 157–160) Exposure tests on steel sheets were carried out and correlated with temp., SO_2, and surface moisture. The results can be used for rating inland exposure sites.

Study on the atmospheric corrosion resistance of low-alloy steels. I. Effect of composition of steel on the atmospheric corrosion resistance. II. Properties of the copper–phosphorus steel plate K. Horikawa, H. Kubota, T. Oosuka, and M. Kotaki (*Tetsu to Hagane*, 1959, **5**, Sept., 1092–1094; 1095–1097) I. The outdoor testing station at Kawasaki Works is described and data are given for the loss in wt. of 16 steels (analyses given) in industrial atm. over a three-year period. II. Steels containing approx. 0·4%Cu. and 0·08%P show roughly one-fifth less loss in wt. in industrial atm. over a two-year period.—K.E.J.

Atmospheric corrosion-resistance of electrodeposited nickel and chromium Mond Nickel Co. (*Bibliography No.*59/1, 1959, July, pp.36).

Estimating corrosion rates in water W. Stumm (*Ind. Eng. Chem.*, 1959, **51,** Dec., 1487–1490) Measurement and interpretation of polarization resistance is described.

Research on the resistance of tinplate to corrosion S. C. Britton and K. Bright (*Chem. Ind.*, 1960, Feb. 13, 158–163) Work of the Tin Research Institute is outlined.

Experiments on the corrosion of nickel-plated steel sheet in the open air E. Raub, F. Elser, and G. Schmidt (*Metalloberfläche*, 1959, **13,** Oct., 330–336) It was found under the conditions examined that Ni plating with an intermediate Cu layer (provided that the latter was polished) gave better resistance to atmospheric corrosion than a single layer of Ni; also OH steel when plated was somewhat more resistant to corrosion than basic Bessemer steel.

The corrosion of austenitic stainless steels under heat transfer in high temperature water J. N. Wanklyn and D. Jones (*J. Nuclear Mat.*, 1959, **2,** July, 154–173) Oxygen-free water at 280° and pH 11 (by addition of KOH) was used. Corrosion was not greatly increased by heat transfer but intercrystalline corrosion could occur in crevices. The concentration of KOH has influence especially when it is high, and it may be more aggressive than NaOH.

Fretting corrosion of metals R. T. Allsop (*Metallurgia*, 1959, **60,** Sept., 87–92) Factors influencing fretting corrosion are considered, and methods of reducing or eliminating fretting damage are discussed with reference to the literature (30 refs).

Underwater corrosion of steel International Paints (*Propeller*, 1960, (4), pp.4) A discussion of conditions for corrosion and control with a note on proprietary primers.

Corrosion studies at the National Chemical Laboratory; immersed corrosion investigations—2 (*Corros. Prev.*, 1959, **6,** Dec., vi–viii) An extract from the 1958 report of the Laboratory, dealing with the investigation of corrosion of metals in flowing water, and of mild steel tubes in Scotch marine boilers, and electrochemical tests on immersed paint films on steel.

Corrosion factors affecting the choice of stainless steels for chemical plant H. T. Shirley (*Chem. Ind.*, 1959, Dec. 5, 1953) A report of a lecture.

Selection of alloys for refinery processing equipment A. J. Freedman, G. F. Tisinai, and E. S. Troscinski (*Corrosion*, 1960, **16** Jan., 19t–25t) The need for a corrosion testing programme and the metallurgical requirements for structural materials for refinery processing equipment is stressed.—T.G.

Corrosion as a design problem S. G. Clarke (*Chem. Ind.*, 1960, Jan. 23, 84–89) A lecture.

Influence of the silicon in cast iron on corrosive wear R. Graham, O. S. Prado, N. H. Collins, E. A. Brandes, and H. K. Farmery (*Proc. Inst. Mech. Eng., preprint*, 1960, pp.10) Cylinder liners for Diesel engines and specially cast specimens were examined for corrosion and erosion, and engine tests were also carried out.

Corrosion cracking of stainless low carbon steel B. V. Linchevskii and V. V. Romanov (*Doklady AN*, 1959, **129,** (5), 1020–1023) The rate of corrosion of Kh18N9 steel in boiling HNO_3 and its tendency to intercrystalline corrosion after testing in a solution of $CuSO_4 + H_2SO_4$ with copper chippings have been studied in a series of experiments and the results are tabulated. The findings show that a change in the carbon content from 0·1 to 0·0146% has no appreciable effect on the corrosion resistance of this steel and to cracking in a 42% solution of $MgCl_2$.

Resistance of certain steels to cavitation corrosion M. G. Timerbulatov and G. I. Babushkina (*Metallov. Term. Obra. Mat.*, 1959, (8), 5–12) Structural, tool, stainless Cr and Cr–Ni austenitic steels have been investigated for the effect of hardness on the losses of material due to cavitation and on the depth of erosion. Certain tool steels, which were tempered at a low heat after quenching for a hardness $>330H_B$ show great resistance to cavitational erosion.

The effect of polarization on corrosion cracking of metals V. V. Romanov (*Doklady AN*, 1959, **129,** (4), 873–875) Quantitative investigations of the effect of polarization on the corrosion cracking of Cr–Ni and carbon steels, of Al and Mg alloys, of brass and of an alloy of 63% Mn + 18%Cu + 20%Ni made it possible to conclude that independent of the inter- or intracrystalline character of corrosion cracks the individual properties of the medium and of the metal affect the typical shape of the polarization curve against temp. and initial voltages. This dependence reveals a relation between the velocity of corrosion cracking and the density of the polarizing current.

The application of Faraday's Law to the study of the anodic dissolution of metals M. Froment (*Corros. Anticorros.*, 1959, **7,** Feb., 46–55) Optical methods, and the dissolution of metals in the form of stable ions are treated in this section.

Investigation of the kinetics of the electrode processes occurring in metals dissolving in acids. II. Nickel in sulphuric acid V. I. Kravtsov and Hsi-bin Hsan (*Vestnik Leningrad Univ.*, 1959, (22), *Seriya Fiz. i Khim.*, (4), 81–92) The cathode and anode ϕ log i curves corresponding to the solution of Ni in 10 N H_2SO_4 were obtained from ϕ–t oscillograms. These were photographed with the current switched off and were in satisfactory agreement with equations derived from the theory of slow discharge. Photographed with the current on, they showed discrepancies from the theoretical values due to the concentration of atomic H on the Ni surface.

Anode polarization of high carbon steel in sulphuric acid G. P. Maitak (*Ukrainsk. Khim. Zhur.*, 1959, **25,** (3), 385–391) The anode polarization of high carbon steel in sulphuric acid (from 10–86%) was studied by plotting volt amp. curves for current densities between 0·065 to 100 a/dm². A pronounced hysteresis between the forward and reverse direction of the curves was observed. Curves plotted with increasing current strength consist of two rising branches joined by a plateau corresponding to the limiting current strength conditioned by the passivation; the reverse curves approach the abscissa without jumps in the direction of the second rising branch of the first curves.

Effect of 8-quinolinol additions on the rate of solution of iron and aluminium in hydrochloric acid I. E. Titova (*Zhurn. Fis. Khimii*, 1959, (10), 2173–2177) The dissolution of Fe and Al in 0·51; 1·06; 1·69; and 2·28 N HCl with additions of 8-hydroxyquinoline in a range from 0·001–0·04 M has been studied. The oxine was found to be an inhibitor for the solution of iron and its action increased with increasing concentrations. As regards Al however, these additions could either accelerate or retard dissolution according to concentration. This behaviour may be explained by the chemical properties of the oxine and by those of the iron and Al surfaces.

Rust preventive properties of nonylphenoxyacetic acid in some petroleum products R. M. Pines and J. D. Spivack (*Corrosion*, 1960, **16,** Jan., 16–19) Laboratory tests are described which indicate nonylphenoxyacetic acid to be an effective corrosion inhibitor for ferrous metals in a variety of petroleum-base fluids. The inhibitor effect was found for aqueous and non-aqueous systems under static and dynamic conditions.—T.G.

Modern methods are winning battle against corrosion (*Ind. Fin.*, 1959, **11,** Nov., 26–27) Details are given of the Corrosion Section's Exhibition at the Oct., 1959 Open Days at the BISRA Battersea Laboratories. Recent advances in the prevention of corrosion in air, seawater, and soil are summarized, and improved testing and control apparatus is described.

Cathodic protection of naval ships and structures (*Annual Rept. Defence Standards Laboratory, Commonwealth of Australia for the year ended June*, 1959, 50) The results obtained in the last year in both laboratory and ship trials are described. **The preservation of steel** (51–52) The effects of various pretreatments on steel plates, subsequently weathered for up to 10 months were studied, both in terms of preservation of the steel, and of the performance of finishing schemes on freshly applied priming coats as compared with weathered priming coats. **Electrode potential studies of pure iron in solutions of sodium nitrite** (52) The main pattern of the behaviour of iron in a $NaNO_2$ solution has been determined, but an acceptable level of reproducibility has not yet been attained. The method of study is described.

Corrosion protection of steel structures by painting: surface preparation—antirust paints E. Wahl (*VDIZ*, 1960, **102,** April 21, 467–469) The most important factor in corrosion protection of steel structures is the life of the coating, and this depends on the pretreatment of the steel and on the use of the correct paint for the individual purpose. The methods of pretreatment are reviewed very briefly and the principles for the selection of the correct paint are set out. The cost of pretreatment is about 40% of the total cost.—T.G.

Chromate and phosphate treatments for corrosion protection D. J. Fishlock (*Corr. Prev. Control*, 1959, **6,** Aug., 42–46, 51) A review of treatments of ferrous and non-ferrous metals.

Combating the secondary oxidation of steel Yu. T. Lukashevich-Duvanova, A. P. Pronov, B. N. Ladyzhenskii, and L. M. Belova (*Stal'*, 1960, (4), 299–302) Protection of 40 KhN basic steel during pouring in the absence of a slag phase can be effected by securing the presence of finely divided slag particles. These are present in acid OH steel and the process of introducing them into basic steel has now been tried. Small ingots were poured, but with large ingots deoxidizers (especially Si—Ca) should be added to the furnace, not to the ladle, and when the inclusions produced are about as numerous as in acid steel protection is satisfactory.

Greases to prevent fretting corrosion N. A. Scarlett (*Engineering*, 1960, **189,** March 25, 424–425) An account of research by the Shell Co.

Internal cathodic protection: 5—further problems involving closed rectangular boxes: the use of live anodes C. Plumpton and C. Wilson (*Corros. Prev.*, 1959, **6,** Dec., 34–40) Theoretical and practical aspects are discussed.

ANALYSIS

The opening up of iron ores difficult to dissolve in acids O. Quadrat (*Chemia Anal.*, 1959, **4,** (1–2), 405–408) A NaOH–KOH method is described.

Solvent extraction in ferrous analysis W. E. Clarke (*BCIRA J.*, 1959, **7,** Oct., 813–823) The principles of solvent extraction separation are outlined and their application in ferrous analysis is discussed. Some procedures developed in the BCIRA laboratories are described, these techniques may also be applied to ferro-alloys, slags, and the like (35 refs).—T.G.

An expedient for promoting the alkaline fusion of ferrosilicon L. P. Shevach (*Chem.-Analyst*, 1959, **48,** Dec., 104) Filter paper is added to the fusion mixture.

Purity of alloys checked by vacuum fusion analysis (*Steel*, 1959, **145,** Nov., 23 82) A method of analysis is described for checking the gas content of alloys used in steelmaking. Samples are vacuum melted; H_2 and O_2 contents can be measured accurately, but the N_2 measurement is unreliable due to reaction with the metal.—D.L.C.P.

Oxidation of phosphorus during the decomposition of a cast-iron by acid (*Ann. Rep. Defence Standards Lab. Australia*, 1959, June, 10) Though hypophosphorous acid is practically unaffected by dil. HNO_3 even in presence of ferric salts, yet up to 80% of the P content in iron appears as H_3PO_4 on solution. HCl gives up to 25% as oxy-acids, one-third of which is H_3PO_4. Mechanisms are discussed.

Amperometric titrations H. A. Laitinen (*Anal. Chem.*, 1960, **32,** April, 180R–185R) a review for 1958–9 (133 refs).

Volumetric and gravimetric methods for inorganic compounds F. E. Beamish and J. A. Page (*Anal Chem.*, 1960, **32,** April, 249R–261R) A review for 1958–9 (247 refs)

A standard electrode usable up to 1 300°C M. Rey and G. Danner (*Sil. Indust.*, 1960, **25,** Jan., 19–22) The classical standard electrode silver/silver salt has been modified in design so as to render it amenable to use at high temp., e.g. in the determination of polarization curves

during the electrolysis of fused substances, the determination of free energies and other thermodynamic characteristics, etc.—P. F.

A new paper-chromatographic quantitative determination of iron in presence of other ions F. Baykut (*Rev. Fac. Sci. Univ. Istanbul*, 1959, **24,** Jan.–April, 83–86) [In German] The method will determine 5–30 γ.

Polarographic analysis in a medium of fused borax. 3. Polarography of iron, cobalt, and nickel Yu. K. Delimarskii and K. M. Kalabalina (*Ukrainsk. Khim. Zhurn.*, 1959, **24,** (4), 435–439) The polarographic behaviour of iron, cobalt, and nickel oxides in fused borax has been studied by means of solid stationary electrodes. It has been shown that there exists a linear relation between the magnitude of the diffusion current and the concentration of the dissolved oxide. A formula is given for the relation between diffusion current and temp. from which the activation energy may be calculated.

On the determination of hydrogen in steel-making slag M. Imai, T. Nakayama, Hi Oi and T. Emi (*Tetsu to Hagane*, 1959, **45,** Sept., 1080–1082) The arrangement of apparatus and the vac. fusion furnace are described. H_2 contents are given for various OH and electric-furnace slags, and the variation in H_2 in metal and slag during the refining of low-carbon steel is plotted.—K.E.J.

Study on the vacuum method of taking a sample for the determination of hydrogen in molten steel S. Mackawa and Y. Nakagawa (*Tetsu to Hagane*, 1959, **45,** Sept., 1078–1080) The sampling apparatus is described, and the effects of the sampling method and the position of the analysis are discussed.—K.E.J.

Determination of oxygen in steel by the condensation vaporisation method T. Tanoue and S. Inoue (*Tetsu to Hagane*, 1959, **45,** Sept., 1076–1078) The method is described, and reproducibility figures are given for its application to medium-carbon and stainless steels. Results are compared with those from the Sprengel–Orsat method.—K.E.J.

Alcoholic iodine method for the determination of non-metallic inclusions in steel O. Kammori and E. Futamura (*Tetsu to Hagane*, 1959, **45,** Sept., 1072–1074) The method is compared with the Gakushin method for SiO_2, Al_2O_3, FeO, and MnO in three types of low-carbon steel, and is shown to agree well with the vac. fusion method for total oxygen.

The simultaneous determination of carbon and sulphur in steel C. E. A. Shanahan and R. H. Jenkins (*Metallurgia*, 1960, **61** Jan., 43–47) An oxidation method using O_2 at 1280°C is described.

Qualitative detection of cobalt by precipitation chromatography R. N. Golovatyi, V. V. Oshchanovskii, and N. N. Khudyakova (*Ukrainsk. Khim. Zhurn.*, 1959, **24,** (4), 491–494) It has been shown that it has been possible to detect traces of Co^{2+} in the presence of large amounts of Ni, Fe, Cu, Mo, V by precipitation chromatography. The best absorber was found to be 2-nitroso-1-naphthol without carrier, and 'anionite' 'BC' impregnated with a solution of 2-nitroso-1-naphthol. The minimum detectable quantity is 0·06 μg in a drop with a ratio of Co:Ni of 1:50. A colorimetric method for determining Co in Ni salts is also described.

Photometric determination of cobalt in nickel alloys and steel K. L. Cheng and F. J. Warmuth (*Chem.-Analyst*, 1959, **48,** Dec., 96–97) Nitrilotriacetic acid-perborate is used with anion-exchange separation.

Polarographic determination of manganese R. F. Sympson and V. E. Caldwell (*Chemist-Anal.*, 1960, **49,** March, 19–20) The use of ethylenediaminedi-(*o*-hydroxyphenylacetic acid) is described.

Polarographic study of manganese, technetium and rhenium R. Colton, J. Dalziel, W. P. Griffith, and G. Wilkinson (*JCS*, 1960, Jan., 71–78).

Photometric determination of molybdenum, in the presence of tungsten, as the molybdenum (V)-EDTA complex E. Lassner and R. Scharf (*Z. Anal. Chim.*, 1959, **168,** (1), 30–33, reprint) The yellow Mo^{5+}–EDTA chelate can be determined in the range 2–50 mg Mo in 250 ml using a simple filter photometer in the blue region. Relative error is 1–3% average. W does not interfere even with a 500-fold excess.

The complexometric titration of molybdenum-I E. Lassner and R. Scharf (*Z. Anal. Chem.*, 1959, **167,** (2), 114–117, reprint) Mo^{5+} is determined by back-titration of a measured excess of EDTA with Cu soln., with PAN as indicator.

Colour reactions and paper chromatographic studies of yttrium and zirconium G. Kallistratos, A. Pfau, and B. Ossowski (*Anal. Chim. Acta.*, 1960, **22,** Feb., 195–199).

Polarometric determination of zirconium (IV) with tartrazine G. Popa, D. Negoiu, and G. Baiulescu (*Anal. Chim. Acta.*, 1960, **22,** Feb., 200–202) U does not interfere.

Contribution to the chemical analysis of ordinary and special steels C. Vandael (*ATB Mét.*, 1959, **1,** (8), 221–224) Notes on the determination of As, Cu, S, Si, and V.

A study on leaded free-cutting steels. V. Determination of lead in leaded free-cutting steel by the EDTA method Y. Honda and T. Adachi (*Tetsu to Hagane*, 1959, **45,** Sept., 1065–1067) The limiting amounts of Fe^{++}, Cu, Ni, Cr, Mo, Mn, Al, and Ti which produce no interference are determined. **VI. Determination of lead in leaded free-cutting steel by the fluorescent X-ray analyser** T. Takahashi, T. Muragi, Y. Honda, and T. Sue (1067–1068) The apparatus and method are described.—K.E.J.

Colorimetric determination of magnesium in irons A. K. Babko and N. V. Lutokhina (*Ukrainsk. Khim. Zhur.*, 1959, **25,** (2), 226–229) This method of colorimetric determination of Mg in inoculated cast irons is based on the elimination of the iron by extraction with chloroform in the form of a complex with 1-nitroso-2-naphthol, after which Mg is determined in the aqueous phase. The time taken for the analysis is 20–30 min.

Polarographic determination of magnesium in spheroidal graphite cast iron J. Marcinkowska (*Prz. Odlew.*, 1960, **10,** (2), 53–57) The polarographic method for magnesium described depends mainly on the preparation of a suitable solution of the sample to be examined and the precipitation of magnesium hydroxyquinoline. After adding hydrochloric acid to the precipitate the 8-hydroxyquinoline produced is determined polarographically. The details of the analysis are given and the error was found to be 3–5%. The advantage of the method is its speed and accuracy.

Study on the colorimetric determination of tin in iron and steel. I. On the interference of vanadium in the JIS-Cacothelin method, and its removal S. Morita and N. Inoyama (*Tetsu to Hagane*, 1959, **45,** Sept., 1070–1072) Details show serious interference by V in the determination of tin, and means of overcoming this are discussed.—K.E.J.

Emission spectroscopy B. F. Scribner (*Anal. Chem.*, 1960, **32,** April, 229R–237R) A review for 1958–1959 (282 refs).

Accuracy in spectrochemical analysis A. Arrak and A. J. Mitteldorf (*Appl. Spectroscopy*, 1959, **13,** (4), 85–91).

A low-voltage spark source for spectrographic analysis P. Erdmann and K. Picard (*Techn. Mitt. Krupp*, 1959, **17,** (3), 146–149) A description and circuit diagram are given of a spark generator which is centrally controlled by a multivibrator impulse transmitter, and in which the frequency can be regulated continuously between 1 and 200 Hz. The source has similar properties to that developed by Bunge (*Spectrochim. Acta*, 1955–56, **7,** 139–140) but has simplified electronic control, and is capable of supplying relatively large electric discharges.

Application of the quantometer to production control in steel making Y. Honda and K. Mizutani (*Tetsu to Hagane*, 1959, **45,** Sept., 1063–1065) The method is based on the spectral lines of Fe, Si, Mn, Cu, Ni, Cr, and Mo.—K.E.J.

Spectrographic determination of carbon in low alloy steels and in heat resistant alloys of the vitallium type F. Malamand (*Recherche Aeronautique*, 1950, (15), 39–43; *Translation 450 of Royal Aircraft Establishment*, 1954, Feb., pp.10).

Stylometer determination of sulphur in certain types of steels and welding joints A. S. Dem'yanchuk (*Zavods. Lab.*, 1959, (12), 82–84) A new method is described for determining sulphur in steels and welding joints by means of the standard stylometer ST–7. The arrangement is described and the analytical lines are S 5453·88 Å with the comparison line of nitrogen N 5462·62 Å. The experimental error is ±4%.

Spectrophotometric determination of molybdenum with 9-methyl-2:3:7-trihydroxy-6-fluorone A. K. Majumdar and C. P. Savarnar (*Anal. Chim. Acta*, 1960, **22,** Feb., 158–162).

The simultaneous spectrophotometric determination of copper and nickel in low-alloy steels M. Freegarde and B. Jones (*Analyst*, 1959, **84,** Dec., 716–719) Bis(*cyclo*hexanone-oxalyl dihydrazone is used for Cu and Na_2 ethylbis-(5-tetrazolylazo)acetate for Ni.

Spectrochemical determination of small amounts of aluminium in steel Z. Takao and S. Miyosi (*Tetsu to Hagane*, 1959, **45,** Sept., 1068–1070) Analytical results are given for synthetic standard and practice samples containing Al between 0·05 and 0·30%.—K.E.J.

Determination of small amounts of special elements in cast iron. II. Spectrophotometric determination of small amounts of zinc in cast iron K. Ota (*Nippon Kinzoku*, 1959, **23,** March, 160–164) Bricker's double extraction method is modified and simplified, in respect of the washing of the Zn dithizonate and the use of benzene as solvent for dithizone. The method is suitable for determining 0·0002–0·04%Zn in cast iron and mild steel; other elements (except >0·4%Cu) do not interfere (14 refs).

A solution microtechnique for the spectrochemical analysis of oxide inclusions in steels S. Meyer and O. C. Koch (**Spectrochimica Acta*, 1959, (8), 549–556) The oxide isolates are fused and sparked on carbon electrodes, with cobalt nitrate as internal standard. SiO_2, Al_2O_3, FeO, MnO, Cr_2O_3, CaO, MgO, and CuO are determined with a mean error of ±5·5%.

Studies on analytical methods for trace elements in metals by using radioactive isotopes. III. Determination of tantalum by means of the isotope dilution method. IV. Determination of aluminium with the calcium-45 type cation exchanger H. Amano (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1959, **A, 11,** Oct., 360–366; 367–374) [In English] III. The determination of Ta in the presence of Nb was investigated using ^{183}Ta. Ta was separated as a tannin ppt. at pH 1·9–2·5 and at Ta/Nb ratios up to 1/50. For ratios of 1/100 or more, repptn. would be needed. The recommended method gives accurate results for Ta in steel (11 refs). IV. Al was determined radiometrically by adjusting the sample solution to pH 3·0–4·0 and treating 'twit' a cation exchanger labelled with ^{45}Ca. The eluted radioactivity was equivalent to the Al^{3+} present. Fe^{III}, Mg, and NH_4 did not interfere. Results for Al in carbon steel agree with photometric results.—K.E.J.

Direct determination of aluminium in iron ores after the simultaneous elimination of iron and titanium by extraction F. G. Zharovskii (*Ukrains. Khim. Zhur.*, 1959, **25,** (2), 245–248) The limits of acidity for the interaction of *N*-benzoylphenylhydroxylamine with the cations of the third and fourth analytical groups have been determined as well as the conditions for the extraction by chloroform of the reaction products. The method is applicable for the separation of Ti and Fe from Al by direct extraction with chloroform.

Determination of gases in metals and alloys (*Annual report of the Defence Standards Laboratory, Commonwealth of Australia for the year ended June*, 1959, 9) The features of an improved apparatus for vacuum fusion gas analysis, and the work being done on N_2 and O_2 in Cr metal with this apparatus, are discussed. **Oxidation of phosphorus during the decomposition of a cast iron by acid** (10) Due to anomalies in the oxidation process, when considered as a normal chemical oxidation, work is proceeding to test the hypothesis that oxidation takes place by energy transfer at active sites on the metal surface during the decomposition process. **Spectrographic analysis** (49–50) The work has been mainly directed towards the extension of facilities and methods for determination of trace elements in metals and alloys. The success achieved in the case of the analysis of Cr is outlined.

Study on the chemical analysis of basic slag. III. Determination of P_2O_5, Cr_2O_3 and V_2O_5 in basic slag S. Wakamatsu (*Tetsu to Hagane*, 1959, **45**, Sept., 1075–1076) Recommended absorptiometric methods are compared with volumetric methods. **Determination of sulphur in basic slag** (Oct., 1164–1168) Two simple methods are described. In one, the sample is heated to 1350–1400°C and a flow of O_2 maintained. The gases (SO_2, SO_3, etc.) are absorbed in H_2O_2–$CuCl_2$ soln., which is titrated with NaOH or (in presence of F^-) $BaCl_2$. In the other the sample is fused with Na_2O_2 and extracted. After adding Fe^{3+} and $HClO_4$ soln., the absorption at 380 mμ is measured (18 refs).—K.E.J.

Determination of sulphur in basic slag by the combustion method. II. Determination of sulphur by induction heating combustion–KIO_3 titration S. Maekawa and Y. Yoneyama (*Tetsu to Hagane*, 1959, **45**, Sept., 1062–1063) Good agreement is found with gravimetric methods for OH, electric furnace, and cupola slags.

Spectrographic analysis of slags Li Chi-so, Chou Si-nun, and Ko I-chen (*Acta Met. Sinica*, 1949, **4**, (2), 169–172) Two methods of spectrographic analysis are explained which can be used to control the production of iron and steel. One is analysis of a solution with carbon electrodes. The slag is fused with borax and then dissolved in 50 cm³ of 30% citric acid. A simple spark is used to take the spectrum. The solution is applied on the flat end surface of a carbon electrode. Si is used as standard. Time of exposure is 60 sec. The other method consists of analysing a solution of the slag in HCl with additions of V as internal standard, exposure 40 sec.

The analysis of waste water from industrial electrolytic processes A. von Krusenstiern and H. Harst (*Metalloberfläche*, 1959, **13**, Oct., 315–316) It is shown that colorimetric methods are to be recommended if the concentrations of substances to be determined are low, particularly if photoelectric methods of measurement are available.

The determination of cyanide in effluents F. R. Russell and N. T. Wilkinson (*Analyst*, 1959, **84**, Dec., 751–754) Cyanide is distilled from a solution containing also thiocyanate and ferro- or ferricyanides and determined colorimetrically with pyridine-benzidine reagent.

Gas analysis A. P. Hobbs (*Anal. Chem.*, 1960, **32**, April, 54R–63R) A review for 1958–1959 (300 refs).

INDUSTRIAL USES AND APPLICATIONS

Materials for springs J. T. Milek (*Mat. Design Eng.*, 1959, **50**, Aug., 115–126) Manual No.151. Carbon steels 1055 and 1060, oil-tempered and hard-drawn 1065 and 1085 music wire, and 1095 high-C; 6150 Cr–V, 9254 Cr–Si, 9260 Si–Mn, 18–4–1 tool steel, 302, 304, 316, 414, and 17–7 PH stainless and non-ferrous materials are reviewed.

Selection of steel types by the Deutsche Bundesbahn for rolling stock and equipment from the DIN specification DIN 17100 Common structural steels Heckner (*Schweissen Schneiden*, 1960, **12**, April, 156) The German railway authority has selected a limited number from the great variety of steels of the German DIN specifications. These steels are listed in a table, together with their fields of application.—T.G.

Materials of construction: iron, carbon steel and alloy steel R. W. Vanderbeck and E. H. Phelps (*Ind. Eng. Chem.*, 1959, **51**, Sept., Part II, 1178–1184) This is the annual review of materials of construction and improvements during the year are noted (107 refs). **Stainless steels and other ferrous alloys** W. A. Luce and J. H. Peacock (1215–1222) (162 refs).

Flat polished steel for auto bumpers (*Met. Prog.*, 1960, **77**, March, 108–111) The Chevrolet, Michigan, installation is described for the abrasive-belt polishing of sheet stock for the manufacture of bumpers.

Achievements to date in light-gauge building construction F. W. Heathcote (*Sheet Met. Ind.*, 1959, **36**, Aug. Sept., 595–600) With a section of protective treatment. **The contribution of research in developing the use of cold-rolled sections in buildings** R. M. Kenedi and A. H. Chilver (601–612, discussion 612–616) Mainly civil engineering aspects. **Problems of design associated with the use of cold-rolled sections for structural purposes** E. Griffin (577–594, 612) Sections are shown and their structural properties are set out and discussed with methods of computation. The use of high-tensile steels in cold-rolled sections is considered and a further section reviews protection against corrosion.

Advancements in steel storage facilities J. O. Jackson (*J. Am. Water Works Assoc.*, 1960, **52**, March, 365–372) Design and advantages of welded steel storage tanks are discussed.

Steels for steam generating plants K. Kreitz (*Brennstoff-Wärme-Kraft*, 1960, **12**, April, 142–144) A review of the latest developments taking place on the steel sector for application in steam generating plants where steam temp. of more than 650°C are now being used. The general trend in steam generation is towards higher temp., and 550–600°C is quite common these days (33 refs).—T.G.

Low-alloy ferritic steels for the manufacture of tubes, forgings and castings for power plant F. Šicha and V. Foldyna (*Hutn. Listy*, 1959, **14**, (12), 1097–1099) Steels and production technology for plant working up to 560°C is discussed with special reference to methods current in Czechoslovakia.—P.F.

New low-alloy steel for large boilers T. M. Slutskaya (*Avtom. Svarka*, 1959, (10), 94–95) 15 G KhN F steel has the following composition: 0·12–0·18%C; 0·3–0·6%Si; 1·25–1·65% Mn; 0·8–1·1%Cr; 0·4–0·8%Ni (when smelted from Orsk–Kalilovsk ores, the Ni content is residual); 0·05–0·2%V; and not more than 0·3%Cu; 0·04%S, and 0·035%P. Mechanical properties, weldability, etc., are tabulated.

The calculation of safety factors for steam boilers in operation P. Bettzieche (*Hutn. Listy*, 1959, **14**, (12), 1169–1171) The problem of safety factors in boiler design is discussed. A nomogram for estimating the useful life of boilers as a function of operating temp. and stress is given.—P.F.

The use of austenitic steels in the new power station at Hüls A. Engl (*Mitteilungen der Vereinigung der Grosskesselbesitzer*, 1958, Aug., 255–264) The types, compositions, dimensions, and design and fabrication of the tube and other parts of austenitic steel used in the plant are described.

Steels for large forgings and castings for steam turbines as used in the V.I. Lenin works Z. Eminger (*Hutn. Listy*, 1959, **14**, (12), 1099–1102) The changes over the last 40 years in the composition of steels used for rotors and casings are surveyed, and present steels and developments are discussed, mainly in relation to practice and plans at the Lenin works (previously Škoda works) in Pilsen.—P.F.

The development of modified 12%Cr steels for steam turbine castings and forgings J. Koutsky (*Hutn. Listy*, 1959, **14**, (12), 1105–1108) The development, properties, and application of a steel containing 12%Cr, 2·3%W, and 0·1–0·3%V is described. Experiences gained with it hitherto show it to be promising for use at 600°C.—P.F.

Selecting materials for high temperature filters H. J. Wheeler jun. (*Mat. Design Eng.*, 1959, **50**, Nov., 108–109) Stainless steels and other materials for use above 350°F are listed.

Efficiency in the use of cold-drawn wire I. N. Nedovizii and A. L. Tarnavskii (*Stal'*, 1960, (3), 280–282) Calculations are presented showing a rolled wire rod to be more economical than the use of drawn wire in building and similar applications.

The modern development of wire ropes D. van de Moortel (*Draht*, 1959, **10**, Sept., 484–488) Mechanical tests on six strand-41 wire Seale–Warrington ropes showed that wire with a tensile strength of 130–140 kg/mm² had the lowest fatigue strength; the best results are to be expected with a tensile strength of 150–160 kg/mm², and for equal strengths galvanized wire gives better results than bright wire. Properties of zinc-coated ropes, and corrosion and lubrication problems are discussed.

Steel wire for wire ropes F. Kohlhase (*Draht*, 1959, **10**, Sept., 471–475) The variation in properties of steel wire manufacture from the upper, middle, and lower portions of the ingot, method of calculation of properties using the 'Draht' universal slide rule, effects of chemical composition on properties, and various aspects of the production of patented and unpatented wire are considered.

Structure, testing and appraisal of high-strength scraping and transport chains for use in mines W. Schaefer (*Glückauf*, 1960, **96**, April 23, 550–562) History, fabrication, use, and testing of chains for scraping machines and for transport are reviewed. It is pointed out that the ultimate tensile strength is no longer the sole criterion for an appraisal of high-strength chains, it is suggested that elongation at fracture and creep should be included, and that these requirements should be incorporated in the German DIN specification. It was found that the suggested elongation values can consistently be maintained by the manufacturers.—T.G.

New uses for stainless steel G. C. Kiefer (*Met. Prog.*, 1960, **77**, April, 120–122) A number of examples of stainless steel articles and fabricating techniques are given.

Household equipment of stainless steel (*Mitteilungen der Forschungsgesellschaft Blechverarbeitung e.V.*, 1959, Nov. 1, 296–297) A note on an exhibition by the information office of Edelstahl 'rostfrei' showing examples of household utensils and other equipment manufactured in various countries.

Stainless steel sections for decorative purposes (*Aciers Fins. Spéc.*, 1960, March, 33–38) Brief reference is made to research on the reduction of weight of metal structures without affecting their strength. Decorative sections in the home and bending tools are cited as examples. The polish on 17%Cr steels is stated to be inferior to that obtained on 18/8 steels.

Sheet materials for high temperature service. Properties and application of 17–7 pH and other precipitation-hardening stainless sheet steel for airframes R. W. White (*ASM compilation from Southwest Metal Congress, May* 1958, 1959, 1–9) The properties are outlined with notes on heat treatment, and uses are indicated. **Ceramic coatings for protection of high temperature materials** N. Cannistraro (25–36) A review of properties, corrosion resistance, and applications. **Electrical strain measurements at high temperatures** J. E. Starr (37–50) Uses of bonded strain gauges are described, also materials used, characteristics, and circuitry. **The cold roll-forming process and as applied to semi-austenitic stainless steels** F. Jacobs (51–68) The various processes, their characteristics, and the methods used and materials formed and their heat treatment are outlined. **Joining and inspection of joints** J. C. Herr (69–74) A small section on the joining and welding of stainless steels is included.

Study of high-boron alloy steels for nuclear reactors M. Imai, T. Nakayama, S. Kinoshita, and J. Hirata (*Tetsu to Hagane*, 1959, **45**, Sept., 1117–1119) Data are given for mechanical properties and structures of Fe–B, 18Cr–10Ni–B and 18Cr–B alloy steels, containing 0·98–4·08, 1·16–2·54, and 1·01–2·61%B respectively.—K.E.J.

The use of sheet in nuclear energy applications F. Pracht (*Mitteilungen der Forschungsgesellschaft Blechverarbeitung e.V.*, 1959, March 20, 66–69) The various applications of steel sheet in atomic power stations are reviewed.

Constructional materials for supersonic aircraft and rockets H. P. van Leeuwen (*Metalen*, 1960, **15**, April 15, 108–114) This instalment discusses the physical and tensile properties and characteristics of the various types of special superstrength and high temperature resistant steel alloys that have been developed for use in the construction of supersonic aircraft and rockets. Various graphs appear in the text and there is a short relevant bibliography.

Present and future of metallic materials used in the manufacture of aircraft and aircraft engines J. A. G. Poggio (*Teen. Met.*, 1959, **15**, Jan.–Feb., 1–60) An exhaustive report (which won the ATEEM 50000 peseta 'M.T. Varela' prize for 1958) describes in detail the present and future needs, uses, and possibilities in aviation of steel and non-ferrous metals and of Mo, Cr, Ta, Ni, W heat-resistant alloys (139 refs).—S.H.-S.

High strength steels for aircraft (*ASM compilation from Southwestern Metal Congress,*

May 1958, 1959, pp.81). **Chromium ultra high strength steels** J. C. Hamaker jun. (1–15) The properties of 5%Cr steel at ordinary and high temperatures, methods of fabrication, surface protection, and typical parts are described. The steel surpasses Ti alloys in the range 400–1000°F (17 refs). **Steel castings in airframes** J. K. Dietz (17–20) A brief survey of the advantages of castings over forgings and weldments. **The use of steel forgings at high strength levels** C. E. Moeller (21–29) Defects, heat treatment, and mechanical properties especially after austempering and in comparison with quenched and tempered specimens are considered. 4340 and 5%Cr steels and others are included. **Materials for hypersonic weapons** J. J. Russ (31–40) Materials including steels are reviewed with special reference to transverse properties. **Machining and fabricating high strength steels** P. S. Kikeli, F. Farago, J. G. Gaul, and R. E. Greenough (41–51) A discussion of various operations with notes on milling and cutting. **Embrittlement of high strength steels** E. P. Klier (53–67) Notch properties and their measurement and hydrogen embrittlement and residual stresses are considered (15 refs). **Acceptance testing of high strength steels** S. Goodman (69–76) SAE 4370, 4330, HS–220, and Hy–Tuf (respectively AMS 6415, 6427, 6407, and 6418) are compared. Their properties are given and methods of testing indicated and discussed. **A practical application of the Holloman–Jaffe parameter to the tempering of Type 410 stainless steel** C. F. Lewis (77–81) Heat treatment of drill steels for oil wells is discussed.

High-strength steels for missiles and spacecraft G. K. Bhat (*Met. Prog.*, 1960, **77,** April, 68A–68B, 151A, 152, 154, 156) Special problems in evaluation for missile production, stress corrosion resistance, and impact strength are among the topics discussed in abstracts of papers presented at a conference on 'High-strength steels for the missile industry' held in San Francisco in Feb. 1960.

Design information on 5 Cr alloy steels for aircraft and missiles F. R. Morral, R. J. Favor, and W. P. Achbach (*PB* 151072, 1959, Aug. 28, pp.48; from *US Res. Rep.*, 1959, **32,** Dec.11, 751) A summary of H–11 type steels. Heat treatment, melting practice, and corrosion protection and fabrication are discussed. High temp. properties are collected and evaluated.

A new steel for aircraft and missile honeycombs A. C. Gilbraith (*ASTM Bull.*, 1959, Sept., 51) An account of the properties of PH 15–7 Mo stainless after various heat treatments and after subzero hardening.

New alloy fills space-age need Universal–Cyclops Steel Corp. (*Iron Age*, 1960, **185,** Feb. 11, 156–157) The properties of 'Unimach UCX2' a new high strength steel for high temp. applications are described. A typical analysis is: 0·39%C; 0·7%Mn, 1·0%Si; 1·10%Cr; 0·25%Mo; 0·15%V; 1·0%Co. Other features are good machinability, ductility, and weldability.—D.L.C.P.

High-speed turbine-driven rotating mirrors: Note on design, construction and performance B. Brixner (*Rev. Sci. Instr.*, 1959, **30,** Nov., 1041–1048) The production of steel mirrors is included.

HISTORICAL

Cast cannon—the 15th to 19th Centuries P. J. Browne (*Found. Trade J.*, 1960, **108,** Feb. 11, 163–165).

The Iron Bridge builders (*Howden Quarterly*, 1959, (41), 6–10) An account of the Darby family, particularly Abraham Darby, and of Tom Paine.

The Midland iron industry in the early eighteenth century B. L. C. Johnson (*250th Anniversary of the successful use of coke in ironmaking, preprint*, pp.9) With special reference to Coalbrookdale. **Coalbrookdale 1709** R. A. Mott (pp.41). **Ironfounding past and present** J. G. Pearce (pp.12). **The contributions of the Darby family to iron making and to the iron industry** A. Raistrick (pp.10).

Cast-iron bells C. J. Robb (*Found. Trade J.*, 1959, **107** Sept. 24, 213) A note on early iron bells.

Blast furnaces since Darby's day (*Engineering*, 1959, **188,** Sept. 25, 248–249) An illustrated review of the evolution of the blast furnace.

The age of iron: commemoration of 250 years of coke in ironmaking (*Iron Coal Trades Rev.*, 1959, **179,** Sept. 4, 233–234) Notes on Coalbrookdale.

The iron industry M. Sanchez de Palacios (*Met. y Electr.*, 1960, **24,** Feb., 84–87) A brief review of methods used from 1600 BC in Upper Mesopotamia, 1300 BC in Egypt, 900 BC Assyria, 800 BC Greece, 1000 AD in Italy and Central Europe, through Bessemer and OH down to present-day Spanish practice.

Who invented the hot blast? D. Murray (*Colvilles Mag.*, 1959, Autumn, 20–23) A note on the circumstances in which J. B. Nielson invented the hot blast.

Smelting in Catalonia—its place in the historic evolution of metallurgy J. A. Riviere (*Tecn. Met.*, 1959, **15,** Jan.–Feb., 61–63) A short survey, with a scale diagram indicating the eras of development of the industry in Catalonia, with reference to the early influence of ores of Corsica, Sardinia, and Elba, the Basque industry and Stückofen.—S.H.-S.

Metallurgical study on ancient iron nails K. Horikawa and Y. Umezawa (*Tetsu to Hagane*, 1959, **45,** Sept., 1121–1123) Data are given for chemical composition, inclusion counts, hardness and appearance of 15 nails made at various times between AD 607 and 1772.—K.E.J.

The casting of grey iron before the introduction of modern materials and methods of manufacture D. Iturrioz (*Met. y Electr.*, 1960, **24,** Feb., 65–72) A review based on an editorial in the *Found. Trade J.* reporting the annual meeting of the Am. Soc. of Grey Iron Founders, with an address by R. T. Thomson and D. F. Caris of Gen. Motors Co. Inspection Dept., followed by a reference to information presented to the International Foundry Congress in Liège by L. Willner of Sveriges Mekanforbund Foundry Dept. Past and present methods of castings are briefly compared. —S.H.-S.

Steelmaking U.S.A. L. F. Reinartz (*J. Met.*, 1959, **11,** July, 444–448) Historical. The development of the electric steelmaking furnace and the LD process are reviewed, and possible future steelmaking developments are discussed (37 refs).

A supplementary commemoration (*Die Fundgrube*, 1959, (9), 195–197) An account of the celebrations of the unveiling of a memorial plaque to Prof. A. Ledebur in Blankenburg, Harz.

Sterling, Ringwood and Greenwood R. W. Shearman and F. W. Starrat (*J. Met.*, 1959, **11,** Sept., 603–609) An historical survey of ironmaking in the Sterling Forest area along the New York–New Jersey border, now being developed as a mining and metallurgical research centre by Union Carbide Corporation (17 refs).

A great French metallurgist: Jules Garnier (*Aciers Fins. Spéc.*, 1960, March, 112–115) A short biography of Garnier's life and work is given.—R.P.

60 Years Bandstahlvereinigung, 1896–1956 (*Pamphlet, publ.* Dusseldorf 1956, pp.48) An account is given of the historical development of the manufacture of strip steel, followed by an historical sketch of the formation and present scope and activities of the Bandstahlvereinigung. Applications of hot- and cold-formed strip are reviewed.

ECONOMICS AND STATISTICS

The steel industry. Foreword (Sir) Cyril Musgrave (**1**) (*Financial Times Supplement*, 1959, Aug. 10, pp.12). **Development plans soundly based** (Sir) Ellis Hunter (I). **Over half Britain's exports based on steel** (II). **Changing styles in demand** E. T. Sara (III). **The market for heavy steel and re-rolled products** T. R. Craig (IV). **The next five years** (IV). **Sheet and tinplate for the car and canning lines** H. F. Spencer (V). **Movements in consumers' stocks** (V). **Developments in special steels** (Sir) Frederick Pickworth (VI). **Main centres of production** (VII). **A wide variety of steel tubes in universal use** A. G. Stewart (VII). **The industry's research task—1: anticipating the designer** W. E. Bardgett and K. J. Irvine (VIII). **2: New methods of production** (VIII). **Steel and the state** L. Chapman (IX). **The prices structure** (X). **Who owns the industry?** B. S. Keeling (XI) **Recruitment and training** W. F. Cartwright (XI). **The long record of good labour relations** (Sir) Lincoln Evans (XII).

Prospects of the steel industry B. S. Keeling (*Metallurgist*, 1960, **1,** May, 108–112) A review with statistics.

Is it possible to supply cheap iron? A criticism of the long term policy of the iron and steel industry S. Ohashi (*Kinzoku*, 1960, **30,** Feb. 15, 1–10) Explanation of the third plan for rationalization with a discussion on estimating long-term demand. To continue the policies of re-investment, the lack of attention to raw materials and techniques which were features of the first two plans, will not be successful.

Shortcomings and achievements of Polish metallurgy in the light of foreign trade S. Kawinski (*Prz. Techn.*, 1960, **81** (9), 10–12) Poland imports 75% of the iron ore used and 7–8% of the coking coal. This has an adverse effect on the foreign trade balance. In the field of partly finished products those of higher quality and price are imported and the cheaper products are exported. However, it is envisaged that in the next five-year plan this position will be improved. In the machine industry there is also a debit in foreign trade, but this is continuously decreasing and it is expected that in five years time there will be a credit.

The main problems in working out wholesale prices for ferrous metals G. N. Khar'kovtsev (*Stal'*, 1960, (3), 264–269) A pricing system under the seven-year plan including transport charges is discussed.

Iron and steel industry of Japan today Japan Iron and Steel Industry Technical Administration Study Team (*Pamphlet*, 1959, Aug., pp.29) Mainly statistical, with brief general sections on techniques, control, training, and research activities.

Evaluation of various types of energy produced in complex processes L. Babický (*Hutn. Listy*, 1959, **14,** (9), 774–777) A simple method of determining distribution of production costs in the individual types of power in complex metallurgical and power plants.

MISCELLANEOUS

Arc decomposition of rhodonite V. Harris, J. D. Holmgren, S. Korman, and C. Sheer (*J. Electrochem. Soc.*, 1959, **106,** Oct., 874–876) A pilot plant for the extraction of Mn is described.

Operational research at work: applying operational research in the steel industry R. H. Collcutt (*Joint BIM/OR Soc. Conference Report*, 1959, Sept. 17, pp.20) A study of the management of the importation of foreign ores the layout of a forge and an OH furnace repair programme are considered with an outline of the solution of each.

Automation European Coal and Steel Community (*High Authority bibliography No.* 19, [1960,] pp.67).

The practical application of industrial television F. Loos (*Technik*, 1959, **14,** Nov., 719–724) Examples are given, with illustrations, of the use of television in rolling mills, welding, mining, the coal industry, power plant, transport, and navigation, and other cases.

Alpha-transformation mechanism of the antiferromagnetic iron sulphide M. Murakami (*Sci. Rep. Tôhoku Univ.*, 1959, **43,** Aug., 53–61) A dilatometric study. A mechanism is proposed. **Anisotropy energy in the paramagnetic region of iron sulphide** (62–66).

Electromagnetic properties and constitutional diagram of the FeO–SiO_2–ZnO system D. M. Chizhikov, V. P. Schastlivyi, and L. I. Blokhina (*Doklady AN*, 1959, **129,** (6), 1353–1355) The region of the pseudo-equilibrium of this ternary diagram was constructed from electromagnetic properties and petrographical analyses. The diagram is not an equilibrium diagram because of the reduction and volatilization of zinc when holding iron alloys at high temp. The consequent changes in the electrical conductivity in relation to composition and temp. are given in graphs.

Some aspects of lighting in the iron and steel industry D. R. M. Nisbet (*JISI*, 1960, **196,** Oct., 207–212) [This issue].

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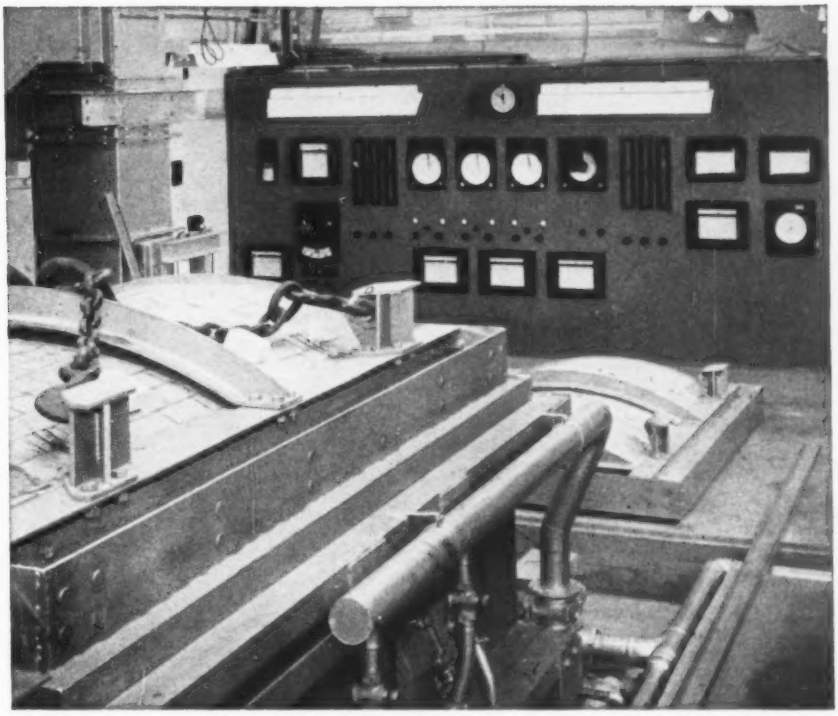

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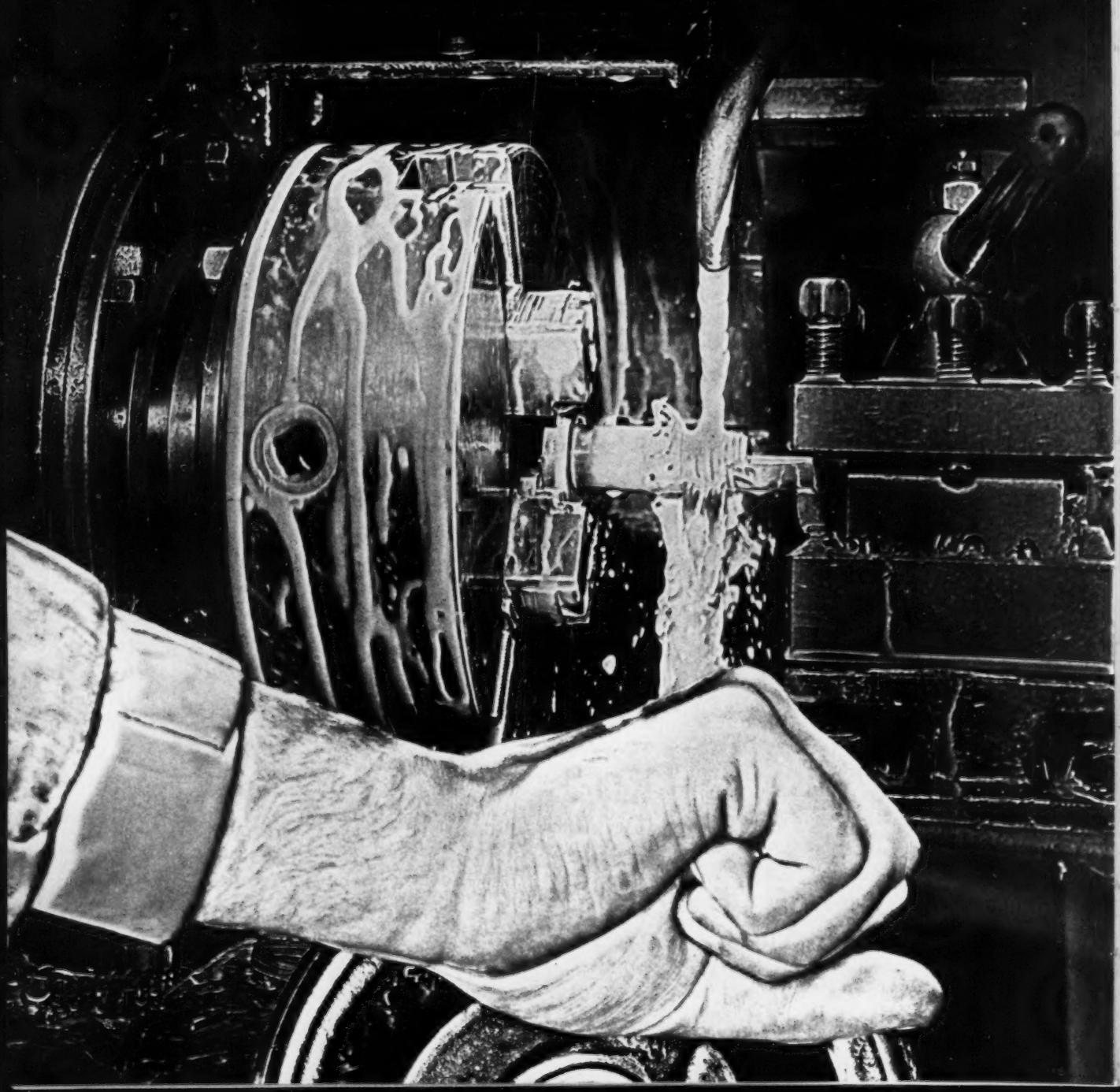


With new Shell Dromus Oils

Most modern soluble cutting oils contain phenolic compounds used as coupling agents between the oil and the emulsifier, for better blending and easier mixing. These phenolic compounds can cause skin irritation, especially where modern high-speed machines are used and the emulsion can concentrate, through the evaporation of water, above the safety level.

Shell research chemists have been working on this problem, which has been causing some concern to Management. After considerable research, Shell Dromus Oils have been reformulated and these new cutting oils now produce bland emulsions, which considerably reduce the risk of skin trouble to operators.

The real difficulty was to find a new coupling agent to replace the phenolic compounds, and Shell finally used what their chemists know as a higher fatty alcohol complex. This solved one problem, but presented another. The new coupling agent was volatile at the high temperatures normally used in blending processes. Further research found a solution to this problem by designing and installing new plant.

The new Dromus Oils are every bit as efficient as before and cost no more. They put Management in the welcome position of being able to minimise working hazards at no extra cost. And machine men need no longer be so worried about skin troubles.

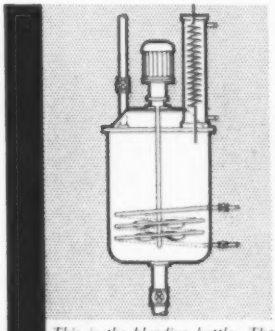
The moral of the story is that Shell research is supremely applicational. The centre at Thornton is always ready to work with even the most specialised sectors of industry to produce the right oil for the job. If you and your organisation have any major lubricating problems, it pays to get in touch with your local supplier of Shell Industrial Lubricants.

The Research Story

Shell chemists in the U.K., in Holland and in the U.S.A., prepared and examined hundreds of experimental soluble oils, and established that certain combinations of fatty alcohols could be used in place of phenolic compounds with no loss of efficiency. They set to work to discover the best combination and developed a higher fatty alcohol complex which fitted exactly. Then they realised that to blend this new coupling agent into soluble oils would require special plant and new blending techniques.

Exhaustive testing of blend stability, emulsion stability, anti-corrosion and machining properties led to selection of the most promising blends. A pilot plant was set up to produce batches of these for use in field trials.

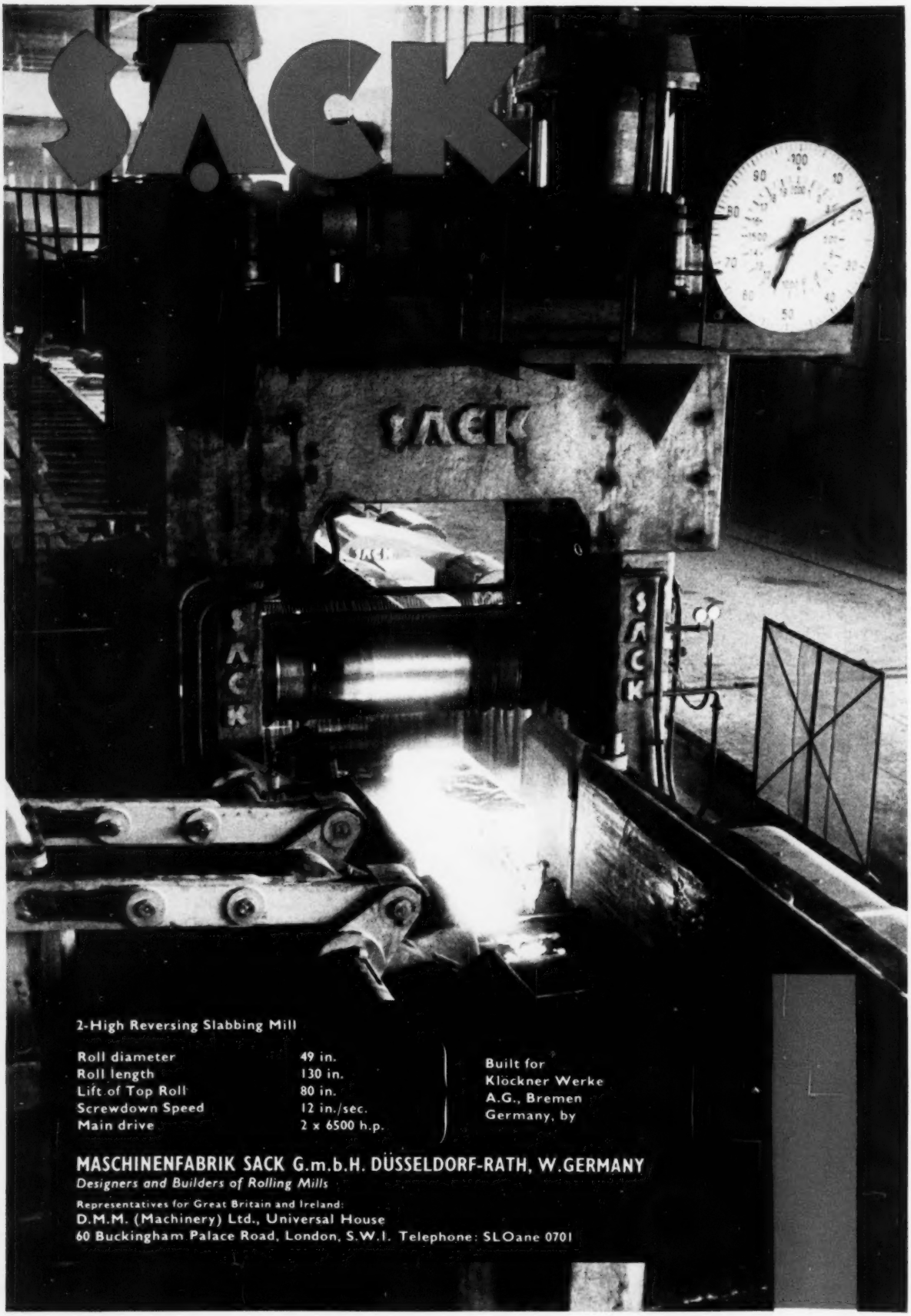
This field testing and final development proceeded for two years whilst production plants were erected at points so chosen as to give the most economical and rapid delivery throughout the United Kingdom.



This is the blending kettle. The reflux condenser beside the stirrer motor prevents the loss of constituents volatile at the blending temperature.

NEW SHELL DROMUS SOLUBLE CUTTING OILS

another proof of Shell leadership in lubrication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There are problems associated with the induction melting of irons and steels which make it imperative for the intending purchaser of an induction furnace to consider all aspects of his production so as to obtain the most suitable furnace for his needs. Even for the same metal, the same furnace will not necessarily be the best type when consideration is given to charge characteristics and melting programme. Having access to the wide range of designs and long experience of Demag, G.W.B. are now in a position to offer the most suitable furnace to suit your own particular project. G.W.B. British built furnaces available for melting irons and steels include:

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- **Channel Type Mains Frequency Furnaces.**

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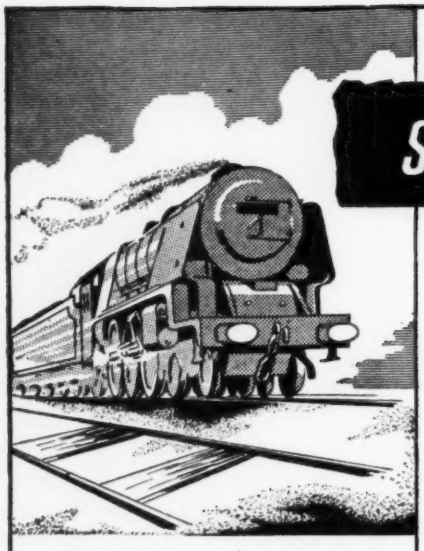

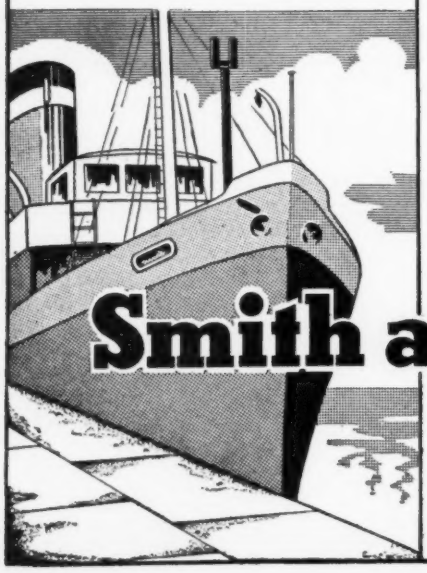

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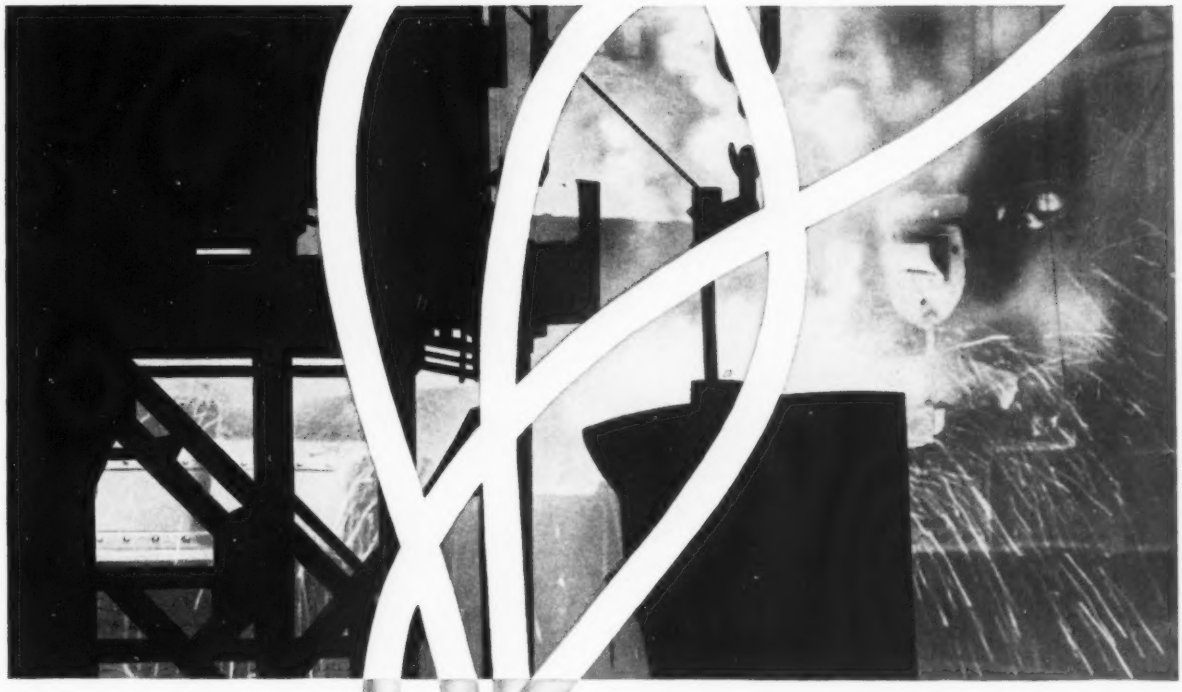

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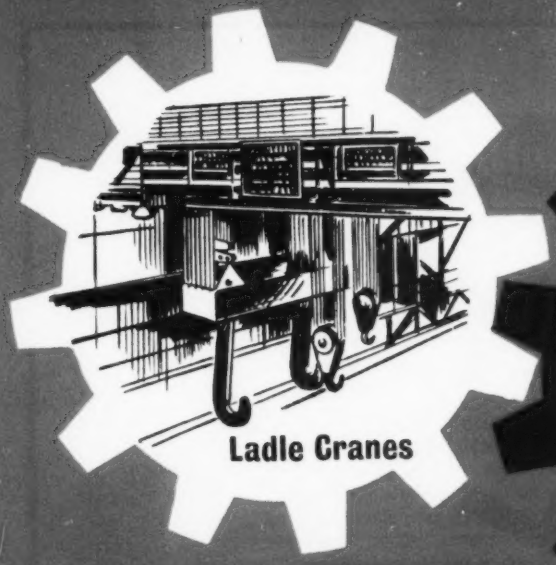

Ladle Cranes

Stripping Cranes

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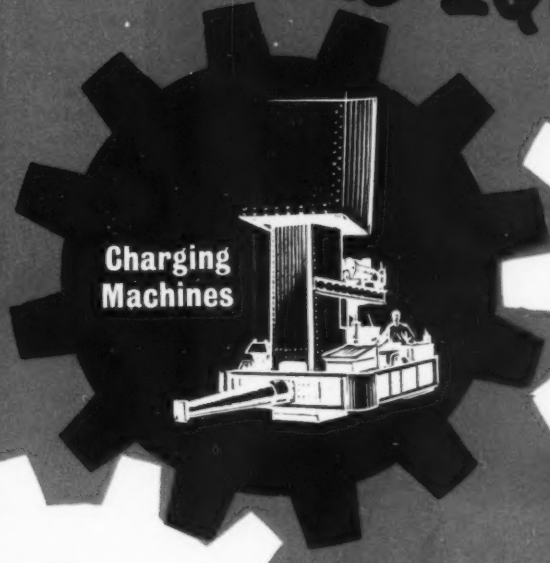
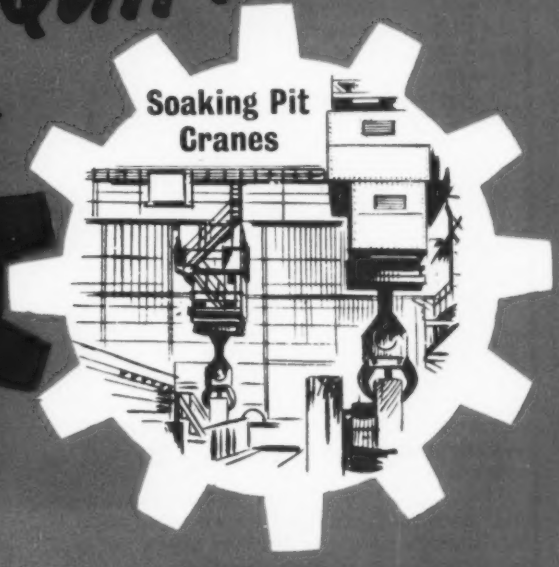


Charging Machines



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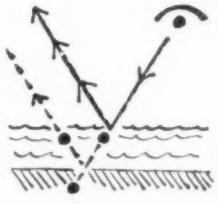
Infra-Red Heating

Infra-red heating is the name given to the method of using in industrial processes radiant heat from heaters specifically designed for that purpose. This differentiates it from methods employing both radiation and convection, as in furnaces, although the effect of the radiation is much the same in both cases.

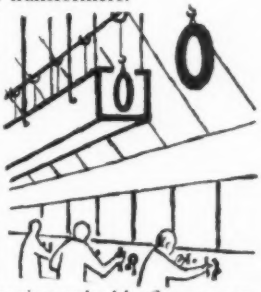
Radiant heat provides a higher rate of heat transfer to bodies not highly reflective than does convection, unless high-velocity forced convection is used. The rate of heating of a body depends upon the intensity of radiation on its surface, the proportion absorbed by the surface and, subsequently, on heat conduction. In heating very thin bodies, for example paint films, the radiation not reflected nor absorbed passes through and is in turn reflected or absorbed by the underlying material. This reflected heat is partly absorbed by the paint film "on the way back". Despite this effect it is an exaggeration to claim that infra-red "heats from the inside out" and it remains largely a surface heating effect except in a few special cases.



Infra-red heaters mainly use metal-sheathed elements in special reflectors, but infra-red lamps with internal reflectors are still used in special cases. Other types use silica-sheathed elements or quartz-sheathed lamps in reflectors. A wide range of heaters is available with simple mounting arrangements to enable heat to be applied quickly and effectively either in a production line or in some odd corner.

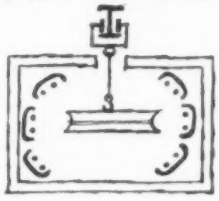
Control of the charge temperature is effected by regulating the heating time and power. In the case of a moving charge, this is done by adjusting the conveyor speed and the number of heaters in circuit, i.e., by adjusting the effective length of the oven. Some of the heaters may be controlled by energy regulators or adjustable transformers.

The advantages of infra-red heating over convection heating include reduced heating times, reduced floor space, lower installation costs, almost instant starting, ability to focus the heat, and flexibility in effective oven length. Infra-red ovens can often be mounted in the roof, saving valuable floor space. Infra-red heating is not generally favourable for heating large bodies of low heat conductivity.



Paint Stoving

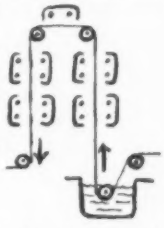
Not all paints are suited to infra-red radiation, but suitable paints are stoved quickly and often give a better finish. Synthetic enamels are the most suitable. Articles successfully stoved include motor car bodies, wheels and other accessories, refrigerator and washing machine cabinets, cylindrical drums and barrels, metal furniture, valve hand wheels, metal toys and a host of other articles. Lacquers and varnishes are being quickly and effectively stoved on jewellery, packing strips, tubes, films and similar products. Special infra-red lamp ovens stove synthetic finishes on wooden furniture and components. In all cases stoving times have been reduced very considerably.



Moisture Extraction

Sheathed element heaters are particularly useful for drying surface water films, their radiation being readily absorbed by water. High rates of drying are possible provided the water vapour can readily disperse. Drying of "contained" water can be effected provided the water can diffuse from the interior at a rate equalling the surface drying rate. With heat-sensitive materials sheathed element heaters may have to be arranged to move away from the material to avoid overheating if the conveyor stops, being interlocked with the conveyor for this purpose. Lamps are sometimes used instead for such materials because of their instantaneous cut-off.

Infra-red is widely used for drying coatings and inks on paper and textile products, bottle seals and labels, clayware and glazes, pigments on leather and many other products.



Plastic Heating

Infra-red heating is widely used in the plastics and allied industries. Typical applications are the preheating of plastic pellets and of sheets prior to moulding, curing of plastic and rubber sheets and belts, heat setting of nylon and similar fabrics, drying and curing of latex films on carpets and many others where a plastic forms the whole or part of the product.

The above are just a few examples of successful applications of infra-red. In many other processes infra-red heaters can often be fitted quickly and cheaply, to save time, labour and floor space and, therefore, money.

For further information, get in touch with your Electricity Board or write direct to the Electrical Development Association, 2 Savoy Hill, London W.C.2. Telephone: TEMple Bar 9434.

Excellent reference books on electricity and productivity (8/6 each, or 9/- post free) are available — "Resistance Heating" is an example; "Induction and Dielectric Heating" is another.

E.D.A. also have available on free loan in the United Kingdom a series of films on the industrial uses of electricity. Ask for a catalogue.

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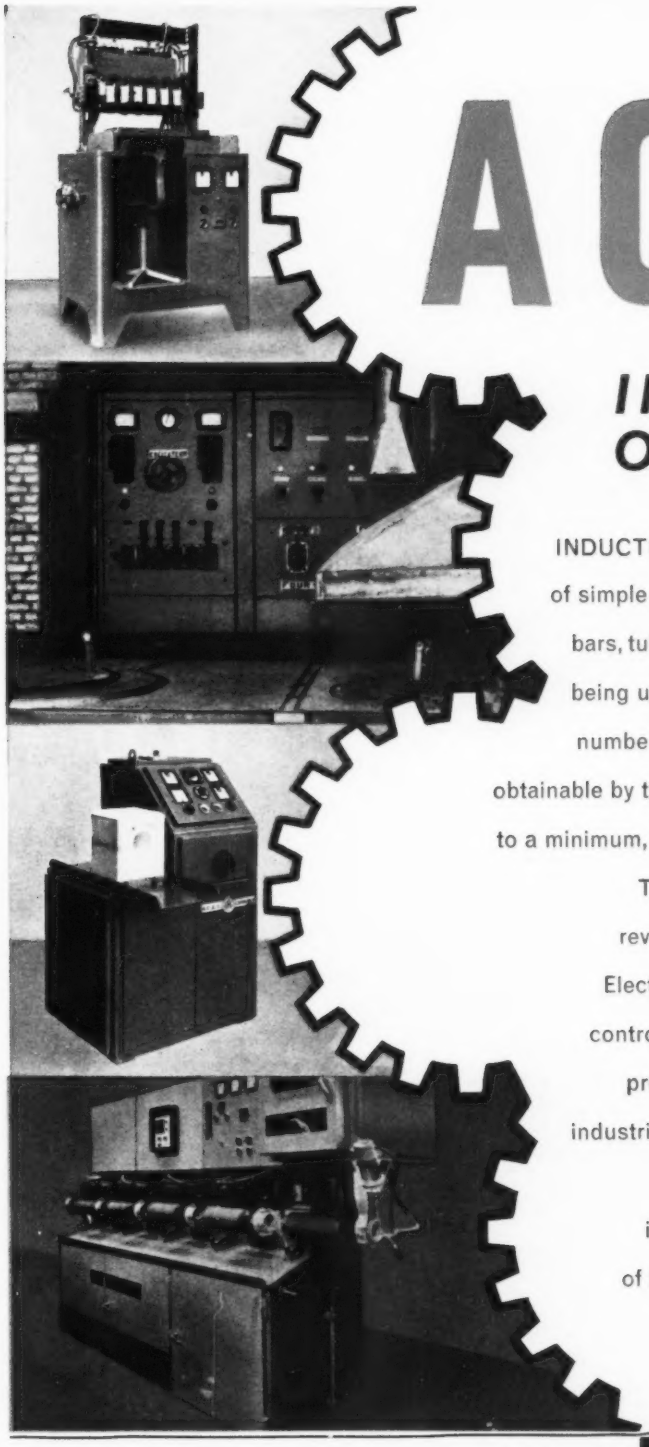

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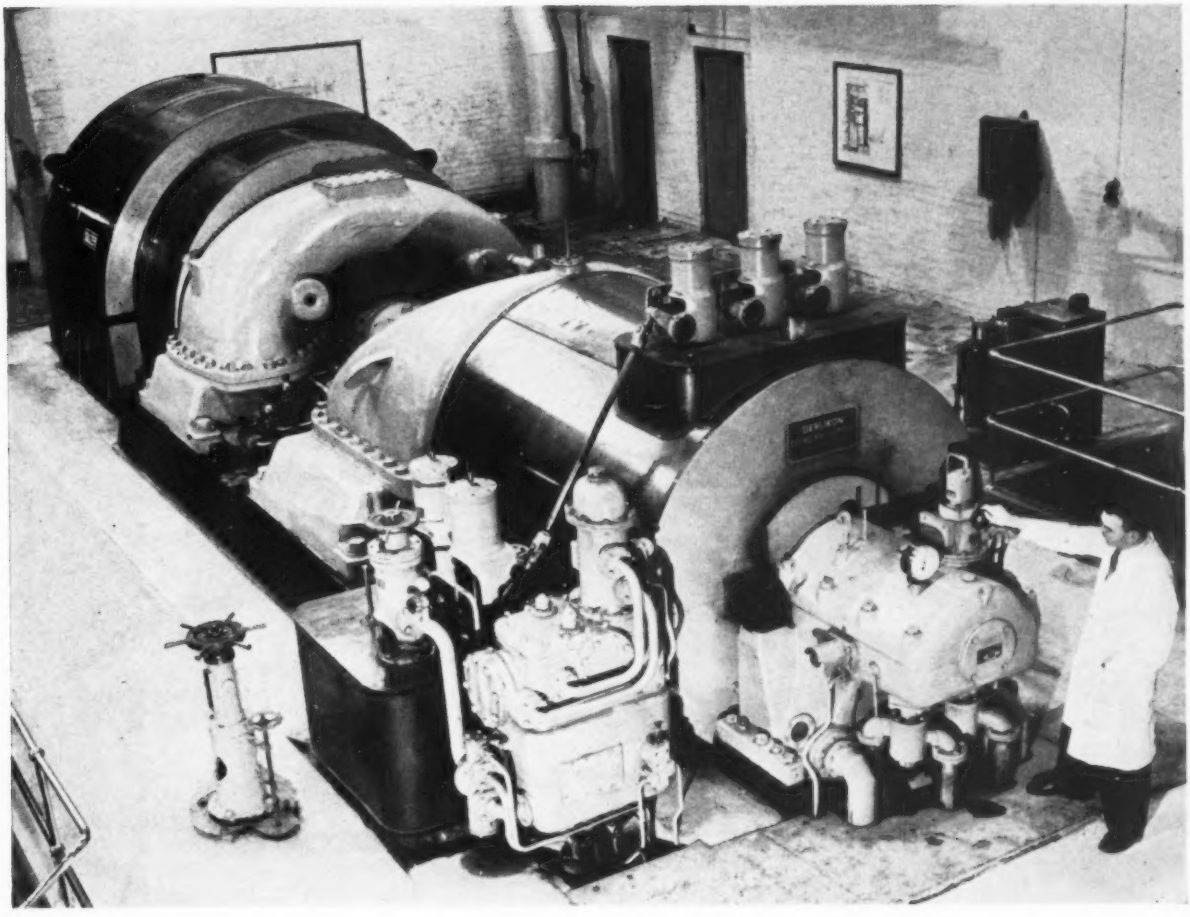

No. 1 Turbo-blower, Margam 'B' Power Station.

Photo courtesy The Steel Co. of Wales

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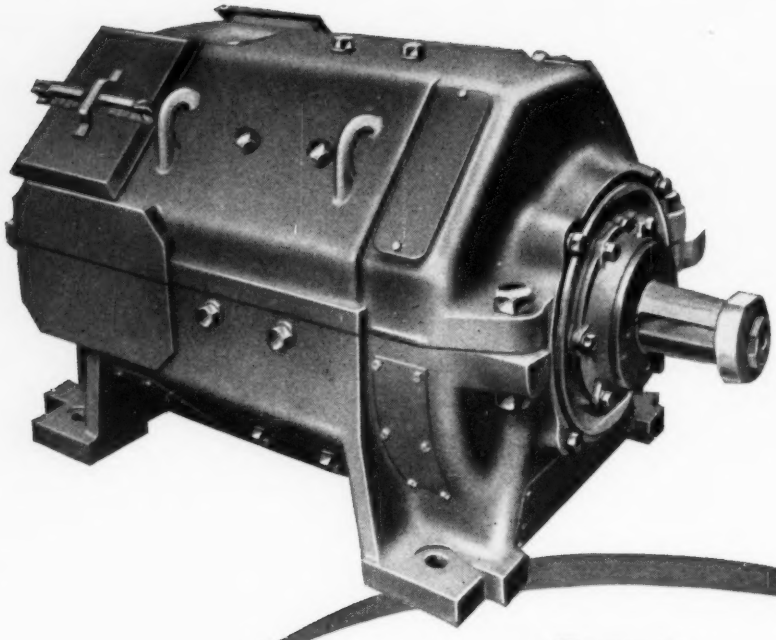

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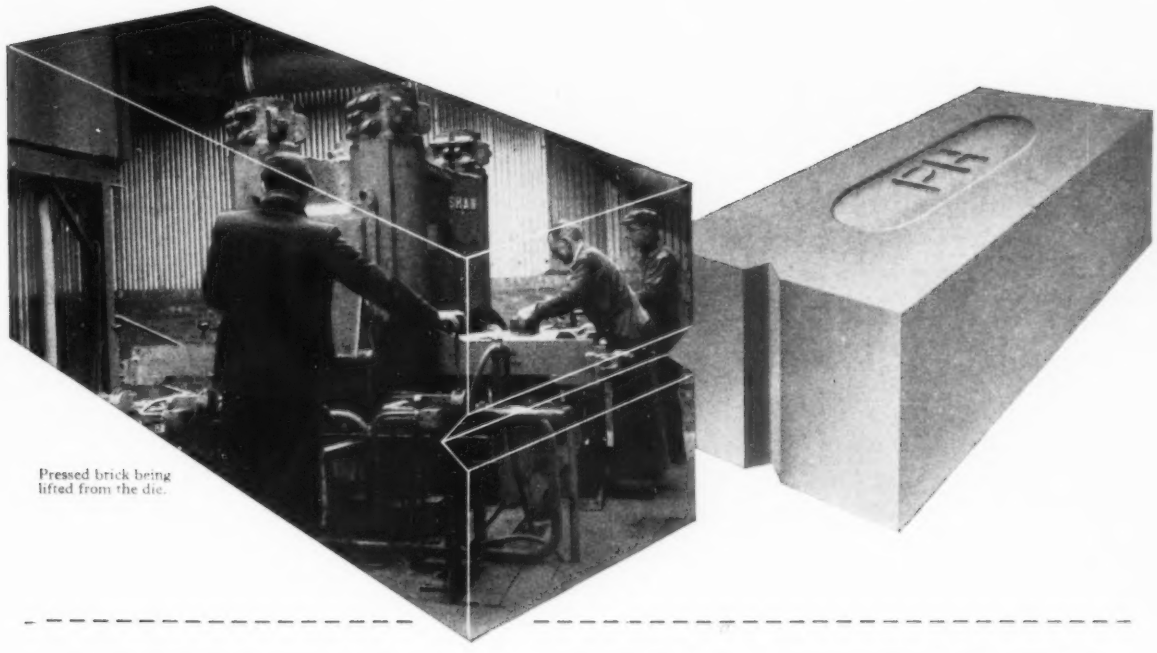

Pressed brick being lifted from the die.

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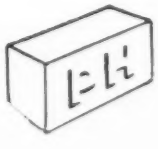

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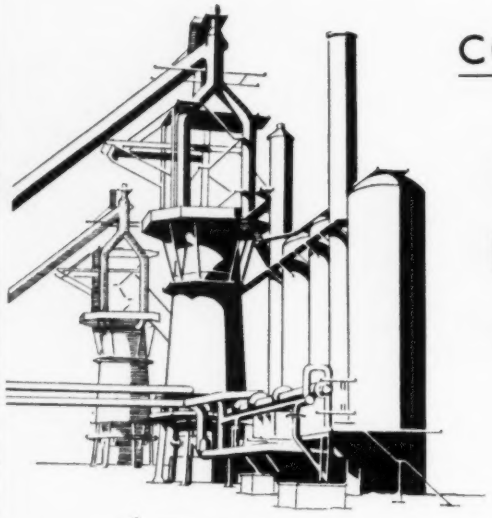

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Special Report

No. 66

Sixth Report of the Corrosion Committee

The corrosion of iron and steel costs industry millions of pounds every year, and research on the best ways of saving this money is of the highest importance. The Sixth Report of the Corrosion Committee, which was founded in 1928 and is now administered by BISRA, has recently appeared as No. 66 in the Institute's Special Report series; it contains an account of the researches carried out during the past twenty years.

The Report covers the corrosion of iron and steel in air, in soil, and in water; it details the results of numerous experiments and recommends methods of protection against the many different forms of corrosion. Reports of tests carried out to ascertain the corrosion rates of different kinds of bare irons and steels all over the world, and of steels protected with paint and/or metal coatings, are given.

Two other interesting researches described in the Report concern the corrosion and protection of steel at elevated temperatures up to 350°C, and marine corrosion. As a result of the latter, several reliable anti-corrosive compositions have been formulated, some of which are in use on H.M. ships and the "Queens".

The Report, which consists of 228 octavo pages, illustrated with 20 line and 21 half-tone figures, and bound in cloth with stiff board covers, costs £3 3*s*. 0*d*. (postage 1*s*. 0*d*.), and may be obtained from:

The Secretary, The Iron and Steel Institute,

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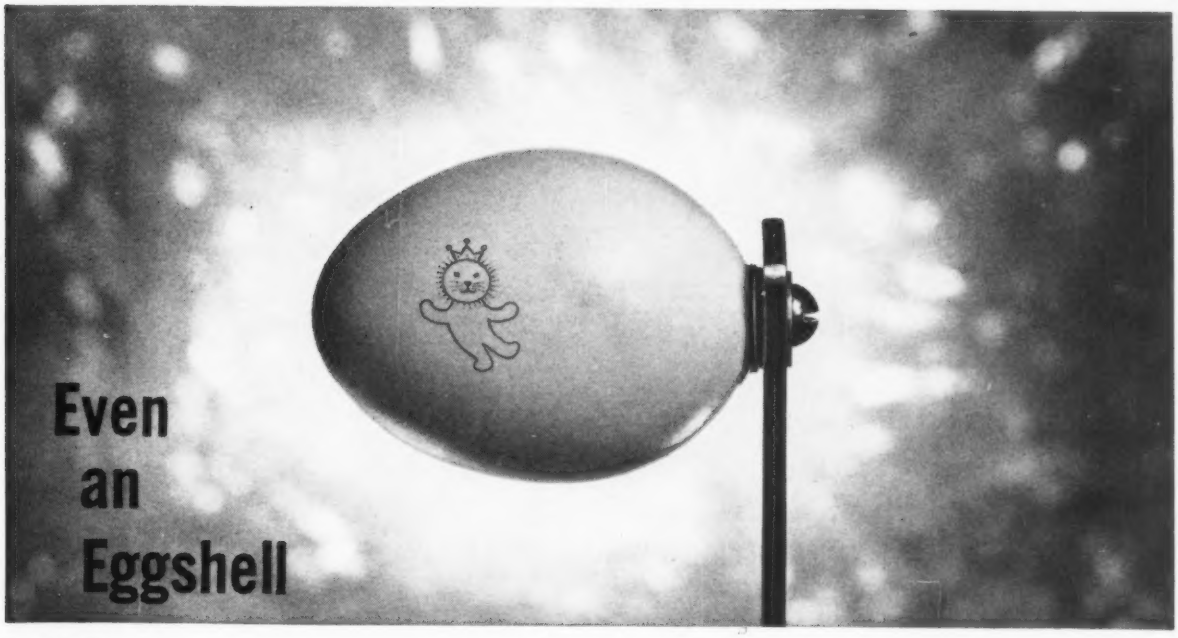

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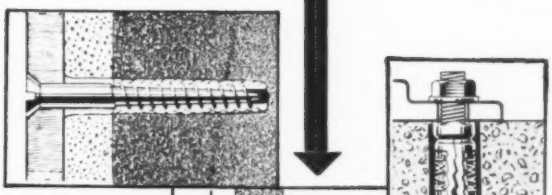

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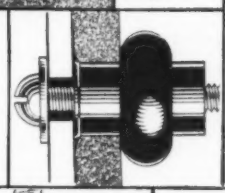

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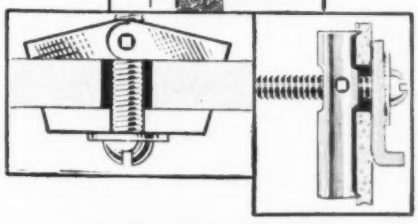

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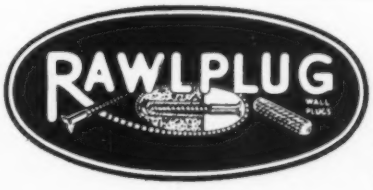

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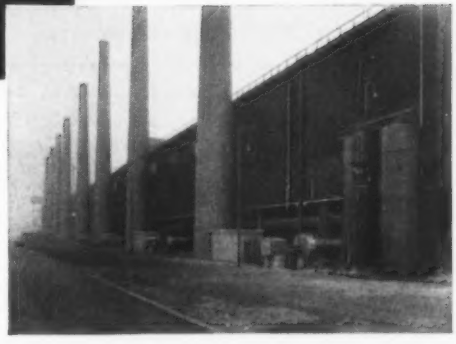

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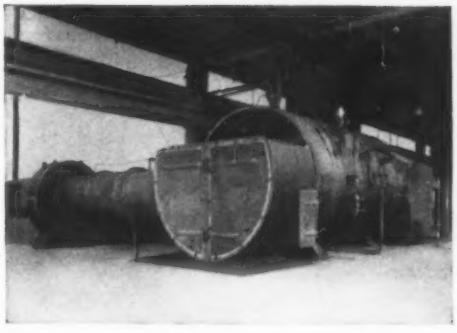

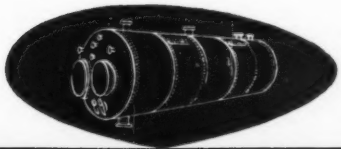

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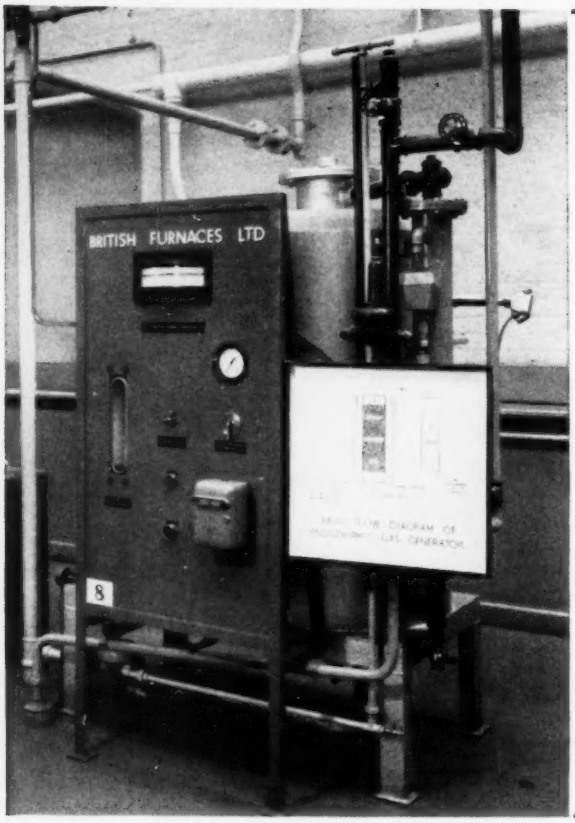

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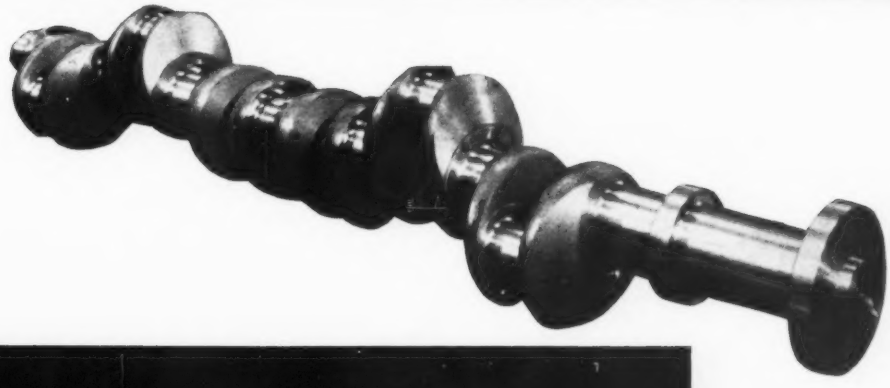

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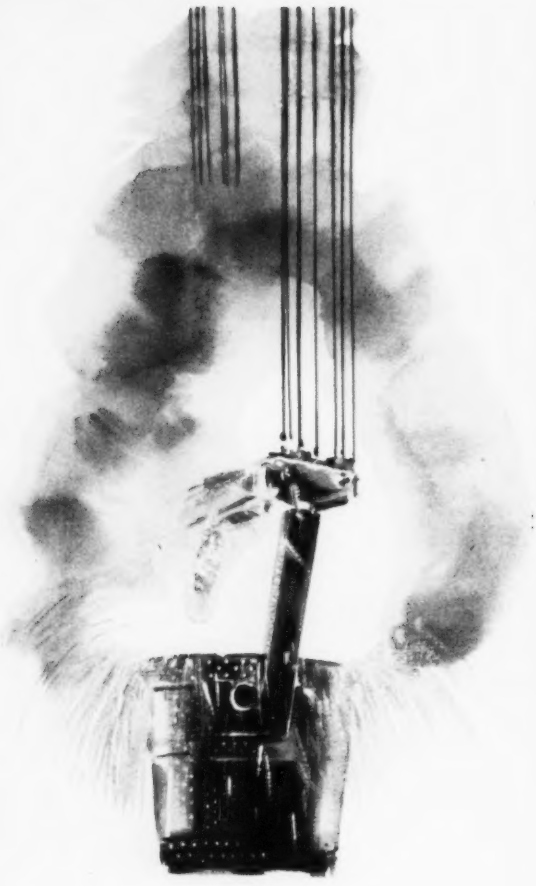

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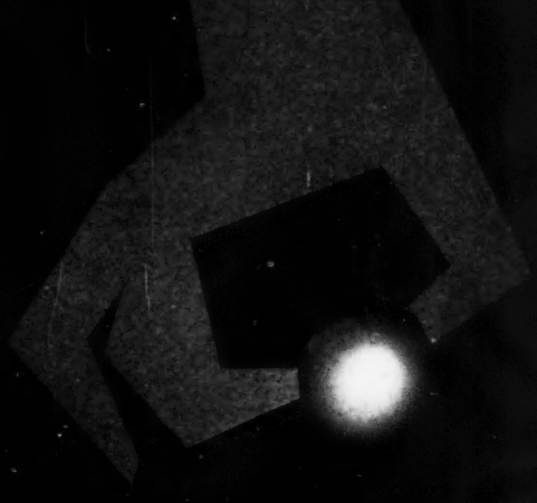

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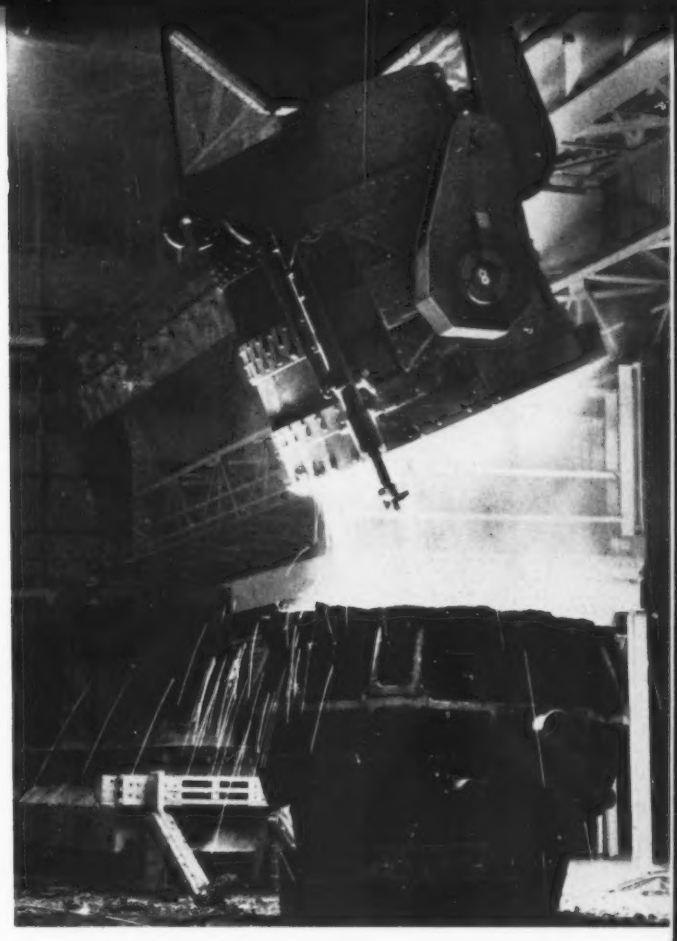

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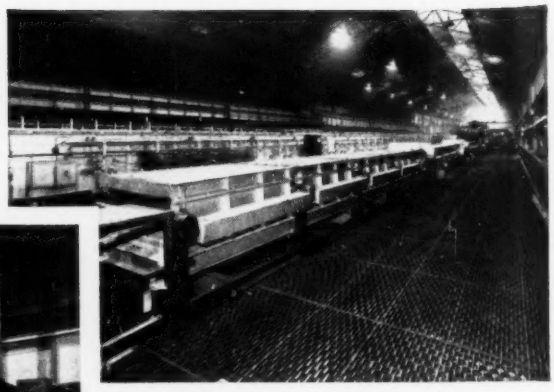

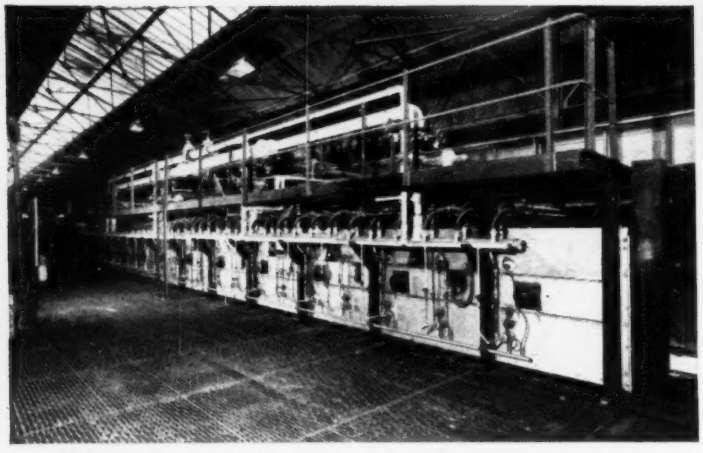

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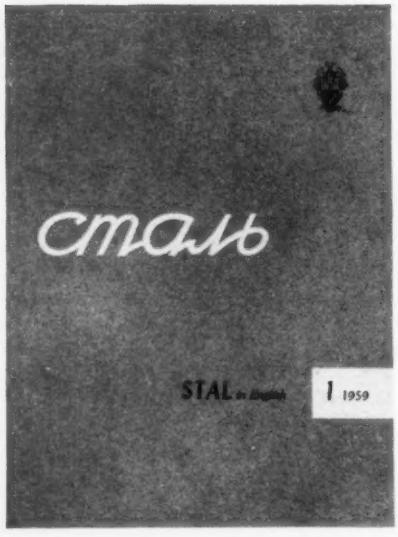

By arrangement with the Department of Scientific and Industrial Research, The Iron and Steel Institute is now publishing monthly a complete English-language version of STAL (*Steel*), the major Russian periodical in the field of iron and steelmaking technology. The developments reported in *Stal in English* are of the greatest interest to everyone connected with iron and steelmaking.

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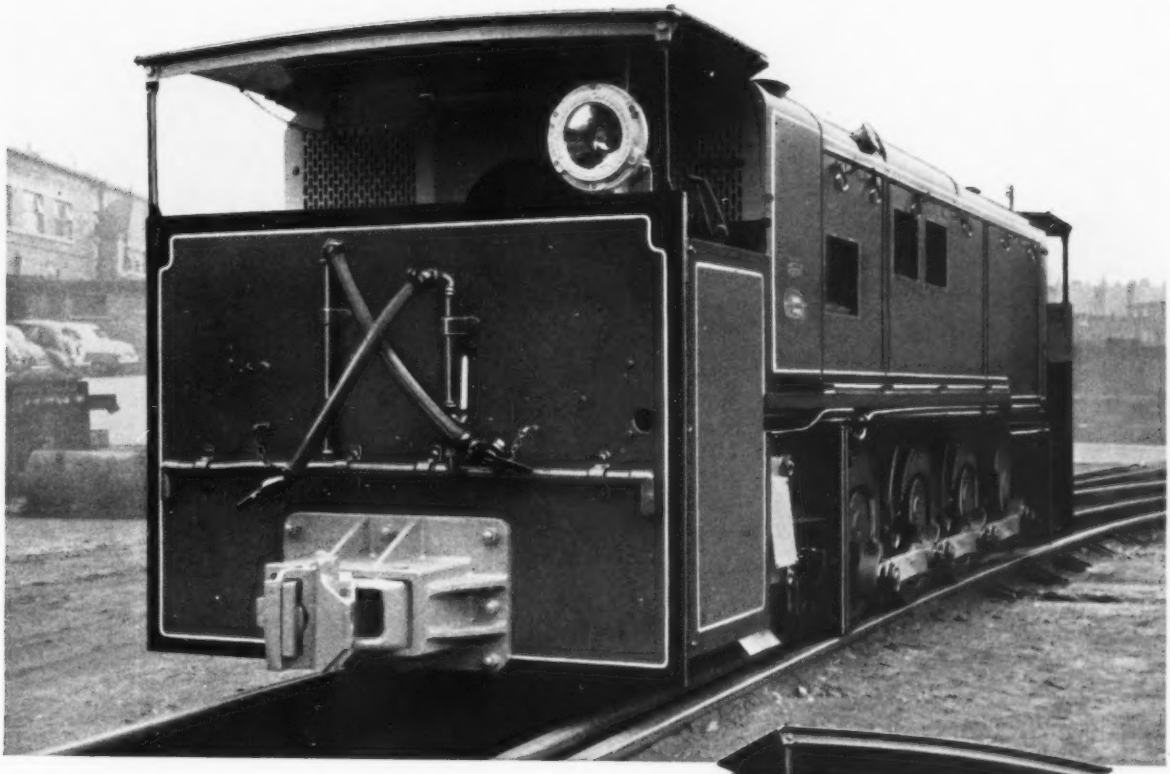

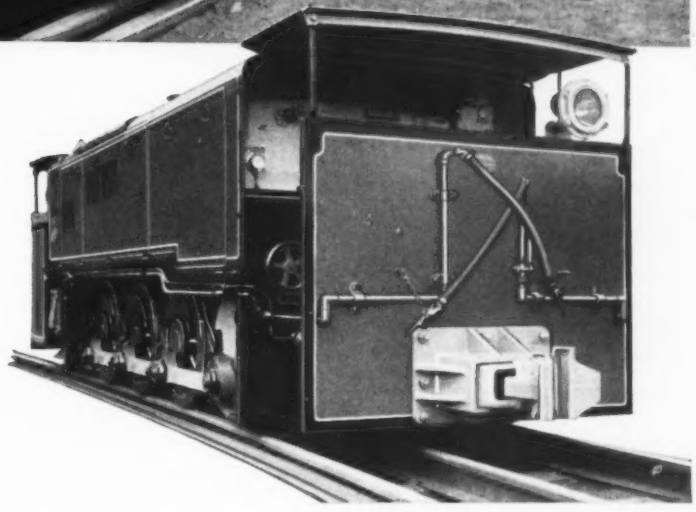

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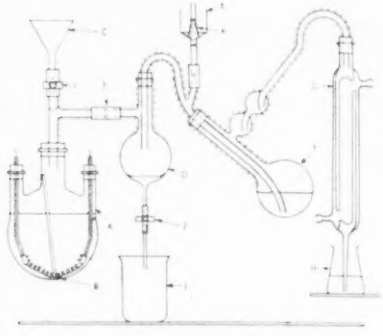

Great interest has always been shown in the role of nitrogen in determining the properties of ferrous alloys. Reliable and accurate methods are therefore necessary for determining the content of this element in alloys.

The establishment of the reproducibility and accuracy of analytical methods to serve as routine or reference procedures for the many alloys now produced is a major undertaking. It was for this reason that the Nitrogen Group of the Steelmaking Division of the British Iron and Steel Research Association was set up eight years ago. The Report of the Group has now been published as No. 62 in The Iron and Steel Institute's Special Report Series. The information it contains will be of inestimable value to all those interested in this subject.

The Report, which comprises 160 quarto pages and a number of illustrations, is bound in cloth with stiff covers, and costs £1 17*s.* 6*d.* (Members £1 5*s.* 0*d.*). It may be obtained from:—

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Index to Advertisers

A.C.E.C. 79
Adamson-Alliance Co. Ltd. . . 75
Adamson, Joseph, and Co. Ltd. . 87
Air Products (Gt. Britain) Ltd. . —
Allen West & Co. Ltd. . . . 13
Armstrong-Whitworth (Metal Industries) Ltd. . . . . —
Arrol, Sir Wm., & Co. Ltd.. . . —
Ashmore, Benson, Pease, & Co. Inside Back Cover
Associated Electrical Industries Ltd. 82

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Bonnybridge Silica & Fireclay Co., Ltd. . . . . —
Bradley & Foster Ltd. . . . —
Brightside Foundry & Engineering Co., Ltd., The 67
British Acheson Graphite Ltd. . . 6
British Electrical Development Association 77
British Furnaces, Ltd. . . . 88
British Iron & Steel Federation, The —
British Paints, Ltd. . . . . —
British Rollmakers Corporation, Ltd. —
British Timken: Division of the Timken Roller Bearing Co. . . —
Brookhirst-Igranic Ltd. . . . 55
Brown Bayley Steels, Ltd.. . . 24
Buell Ltd. . . . . —
Burn Fireclay Co. Ltd., The . . —
Bushing Co. Ltd., The . . . 14

Carblox Ltd. . . . . —
Chemical Construction (G.B.) Ltd. . 50
Colville Group of Companies, The . 52
Consett Iron Co. Ltd. . . . 42
Contactor Switchgear, Ltd. . . 16
Coppee Co. (Gt. Britain) Ltd., The . 22
Cowans Sheldon & Co., Ltd. . . 58
Craven Bros. (Crane Division), Ltd. . —
Crooks, L. E. . . . . —

D.M.M. Machinery, Ltd. 70
Davy & United Engineering Co. Ltd. 64
Demag . . . . . —
Derbyshire Stone Quarries Ltd. . —
Dewhurst & Partner, Ltd. . . 47
Dyson, J. & J., Ltd. Inside Front Cover

Edwards High Vacuum Ltd. . . . —
Electric Resistance Furnace Co. Ltd. —
Electroflo Meters Ltd. Co. . . . —
Engelhard Industries Ltd. . . —
English Electric Co. Ltd. The . . —
English Steel Rolling Mills Corporation, Ltd. 89

Foster, Henry, & Co. Ltd. . . . 74
Foster Instrument Co. Ltd. . . —
Foundry Services Ltd., . . . 10, 11

G.W.B. Furnaces, Ltd. 71
General Electric Co. Ltd., The . . 45
General Refractories, Ltd. . . —
Gibbons Brothers, Ltd. . . . 92
Gibbons (Dudley) Ltd. . . . —
Great Lakes Carbon International Ltd. . . . . —
Guest, Keen, Iron & Steel Co. Ltd. . 76
Guest, Keen & Nettlefolds, Ltd. . 40

Hadfields, Ltd. . . . . —
Head, Wrightson, Iron & Steel Works Engineering, Ltd. . . . . —
High Speed Steel Alloys, Ltd. . . —
Holmes, W. C., & Co. Ltd. . . 57
Holset Engineering Co., Ltd., The . 7
Honeywell-Controls, Ltd. . . . 65
Howden, James, & Co. Ltd. . . —
Hudswell Clarke & Co., Ltd. . . 93
Hunslet Engine Co. Ltd. . . . —
Huntington, Heberlein & Co. Ltd. . 54

Ilford, Limited 66
Industrial & Metallurgical Equipment Ltd. 88
International Combustion Products Ltd. . . . . —
International Construction Co. Ltd., The 32

Johnson's Rolls Ltd. 23
Johnson Matthey & Co. Ltd. . . 31

Kestner Evaporator & Engineering Co. Ltd. 86
Kingscliffe Insulating Products, Ltd. 43

Lamberton & Co. Ltd. Outside Back Cover
Lanarkshire Steel Co. Ltd., The . —
Lancashire Steel Manufacturing Co. Ltd. 56
Laurence, Scott & Electromotors Ltd. —
Lodge-Cottrell Ltd. 26
Loewy Engineering Co. Ltd., The . 15
Lysaght's, John, Scunthorpe Works, Ltd. 2

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Mitchell Engineering Ltd. . . . 8, 9
Mobil Oil Co. Ltd. 41
Mond Nickel Co. Ltd. . . . 39
Moorwood, John M., Ltd. . . . —
Morgan Refractories Ltd. . . . 48
Morgardshammars Mekaniska Verkstads AB . . . . . —
Moxey Ltd. . . . . —

National Coal Board, The . . 18, 19
National Industrial Fuel Efficiency Service . . . . . —
Newell, Ernest, & Co. Ltd. . . —
Newton, Chambers & Co. Ltd. . . 91

Oughtibridge Silica Firebrick Co. Ltd. . 60

Pantin, W., & C. Ltd. 28
Park Gate Iron & Steel Co. Ltd. . —
Parsons, C. A., & Co. Ltd.. . . —
Pearson, E. J. & J., Ltd. . . . 90
Pickford Holland & Co. Ltd. . . 83
Power-Gas Corporation Ltd., The . —
Power Petroleum Co. Ltd., The . —
Press, William, & Son Ltd. . . —
Priest Furnaces Ltd. 29

RN Process Agency . . . . . —
Reavell-Fahie Ltd. . . . 81, 86
Renold Chains Ltd. . . . . —
Rawlplug Co. Ltd., The . . . 85
Research & Control Instruments Ltd. 37
Richardsons Westgarth & Co. Ltd. . 35
Robertson, W. H. A., & Co. Ltd. . 51
Rust-Oleum Corporation . . . —

Salem-Brosius (England) Ltd. . . —
Schloemann AG . . . . . —
Senior Economisers Ltd. . . . 46
Sheepbridge Equipment Ltd. . . 27
Shell-Mex & B.P., Ltd. . . 68, 69
Simon-Carves Ltd. 25
Skefko Ball Bearings Co. Ltd., The . —
Smith, John, (Keighley) Ltd. . . 95
Smith & McLean Ltd. . . . 72
Smith, Thomas, & Sons (Rodley) Ltd. —
Solano Air & Heat Systems Ltd. . 5
Spencer-Bonecourt-Clarskon Ltd. . 33
Stanton Ironworks Co. Ltd., The . —
Stedall & Co. Ltd. . . . . —
Steel Co. of Wales, The . . . 34
Stein & Atkinson, Ltd. . . . —
Stein, John G. & Co. Ltd. . . —
Stephens Silica Brick Co. Ltd. . . 80
Stewarts & Lloyds, Ltd. . . . —
Stillite Products Ltd. 38
Strachan & Henshaw Ltd. . . —
Sturtevant Engineering Co. Ltd. . 44
Summers, John, & Sons Ltd. . . 53
Summerson, Thos., & Sons Ltd. . 36
Swann, Ratcliffe & Co. (Brassington) Ltd. . . . . —

Tattersall & Sons Ltd. 84
Thermal Efficiency Ltd. . . . —
Thermal Syndicate Ltd. . . . —
Thomas (Richard) & Baldwins Ltd. . —
Thornton, B., Ltd. 21

Union Carbide Ltd. 20
United Coke & Chemicals Co. Ltd. . —
United Steel Companies Ltd. . . —

Vaughan, Edgar, & Co. Ltd. . . . —
Visco Engineering Co. Ltd. . . 17

Ward, Thos. W., Ltd. 49
Warner & Co. Ltd. 3
Wellman Bibby Co. Ltd. . . . 94
Wellman Smith Owen Engineering Corporation Ltd., The . . . 1
West Hunwick Silica & Firebrick Co. Ltd. . . . . —
West Instruments Ltd. . . . 90
Whitehead Iron & Steel Co. Ltd. . 78
Wickman Ltd. . . . . —
Wild Barfield Electric Furnaces Ltd. —
Willan, G. L., Ltd. . . . . —
Wincott, G. P., Ltd. 30
Woodall-Duckham Construction Co. Ltd. 62
Workington Iron & Steel Co. . . —
Wragg, Thos., & Sons Ltd. . . 59

Yorkshire Engine Co. Ltd. . . . —

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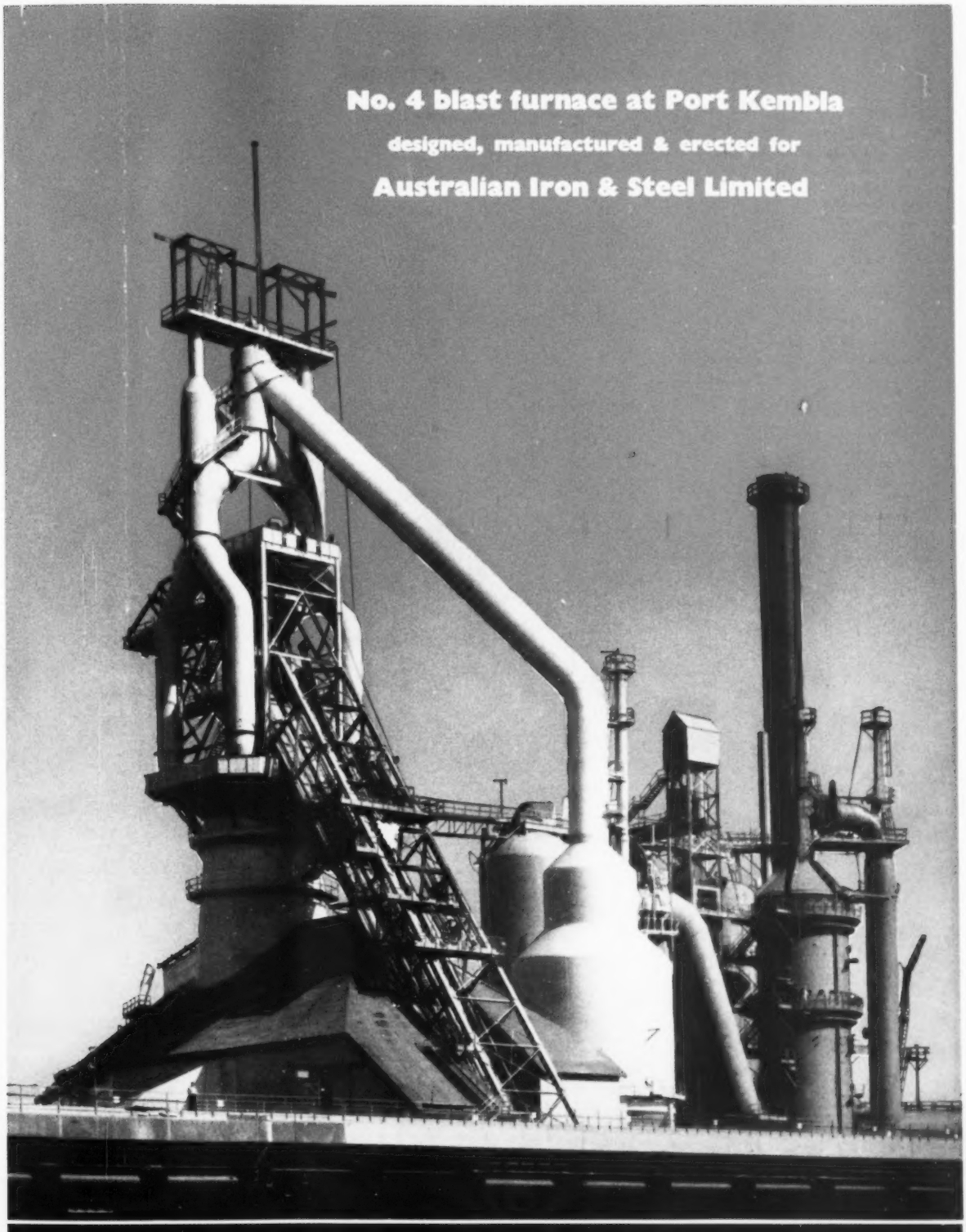

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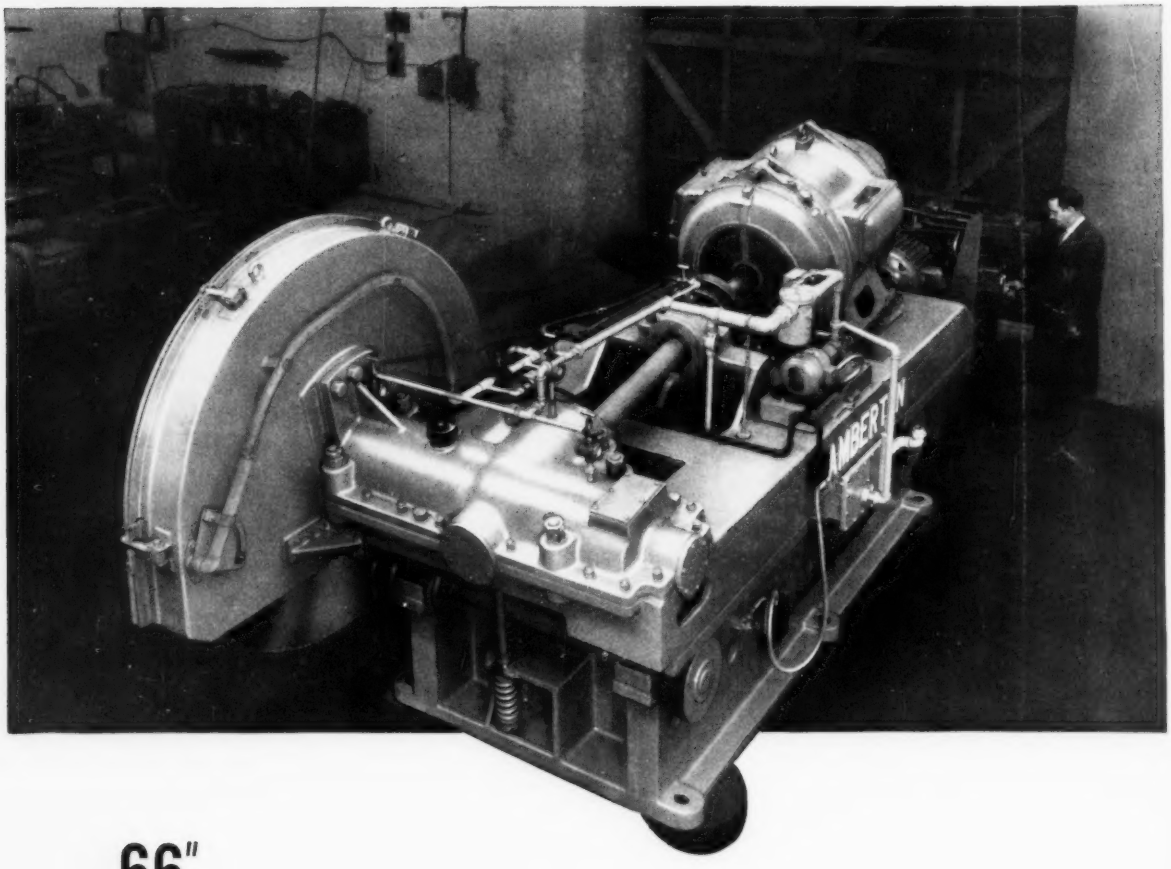

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